THE FRACTURE BEHAVIOUR OF GLASS FABRIC REINFORCED EPOXY CONTAINING STRESS CONCENTRATIONS

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CERTIFICATE

This is to certify that the thesis entitled, "THE FRACTURE BEHAVIOUR OF GLASS FABRIC REINFORCED EPOXY CONTAINING STRESS CONCENTRATIONS", submitted in partial fulfilment of the requirements for the degree of Master of Technology by KINGSHUK BANERJI, is a record of work carried out under our supervision and has not been submitted elsewhere for a degree.

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NOMENCLATURE

a	instantaneous crack length
a_o	initial crack length, characteristic distance for the average stress failure criterion of Whitney-Nuismer
c	initial crack length appearing in the Whitney-Nuismer failure criteria
d_o	characteristic distance for the point stress failure criterion of Whitney-Nuismer
E_L	Young's modulus along a principal fibre direction
E_L', E_L''	primary and secondary Young's modulus along a principal fibre direction respectively.
E_x	Young's modulus along any arbitrary orientation
G_{LT}	shear modulus with respect to principal fibre directions
G_{xy}	shear modulus for any arbitrary orientation
K_R	candidate stress intensity factor
$(K_R)_{ins.}$	critical stress intensity factor at onset of unstable crack growth
P	load
t	thickness of the specimen
V_f, V_m, V_v	volume percentage of fibres, matrix and voids respectively
w	width of the specimen
$Y(a/w)$	calibration factor
ϵ_{cu}	composite ultimate fracture strain
ν_{LT}	Poisson's ratio for straining along a principal fibre direction

ν_{xy}	Poisson's ratio for straining along any arbitrary direction
θ	orientation of a principal fibre direction with respect to the loading axis or the machined notch ($\theta \leqslant 45^{\circ}$)
σ_{cu}	composite ultimate fracture stress
σ_N	notched strength in Whitney-Nuismer failure criteria
σ_o	unnotched strength in Whitney-Nuismer failure criteria, applied stress on the specimen
$\sigma_y(x,o)$	normal stress distribution ahead of a hole or crack in Whitney-Nuismer failure criteria

ABSTRACT

The fracture behaviour of a balanced orthotropic laminate in which the reinforcement was E-glass woven roving mat in epoxy matrix was studied in the Mode I type of loading. Stress concentrations were introduced in the form of a precut notch on one edge of the sample and the load vs. COD obtained. The tests were conducted on samples with varying initial crack lengths and orientation of the principal fibre directions to the loading axis The data was analysed using the R-curve approach by modelling the crack tip damage as a self similar crack extension through a compliance matching procedure. The viability of this approach in treatment of composites of a general nature are discussed. Attempts were also made to ascertain the validity of the stress failure criteria proposed by Whitney and Nuismer for composites containing stress concentrations. It was found that although these criteria remain valid for the systems considered here, the values of the characteristic distances proposed by them were not sacrosanct as they claimed. Tests were also conducted on samples containing centrally situated holes of varying diameters whose principal fibre directions were parallel and perpendicular to the loading axis. However it was found that the Whitney-Nuismer failure criteria did not conform to the data obtained from these samples. The microfracture mechanisms of the composites with their principal fibre directions oriented at different directions to the loading axis were examined and the dominant fracture controlling event proposed for the axial and off-axis orientations.

CHAPTER ONE

INTRODUCTION

1.1 Preview:

Although fibrous composites, both natural and artificial are being used by man since the very beginning of history, they have only assumed the present status in structural and constructional applications after the second World War. Rapid depletion of conventional mineral resources for metals, the oil crunch and the necessity of imparting specific properties to structural members in high performance aircraft, spacecrafts and ballistic missiles led technologists to bear the onus of developing materials with a high strength to weight ratio as well as reasonable toughness. With the advent of aligned fibre reinforced composites a major breakthrough in the structural design concepts was achieved; for the conventional materials such as metals, which are isotropic and homogenous, the design had to be tailored according to the material properties whereas for directionally reinforced composites, the material due to its high orthotropy can be tailored according to the design requirements to a great extent.

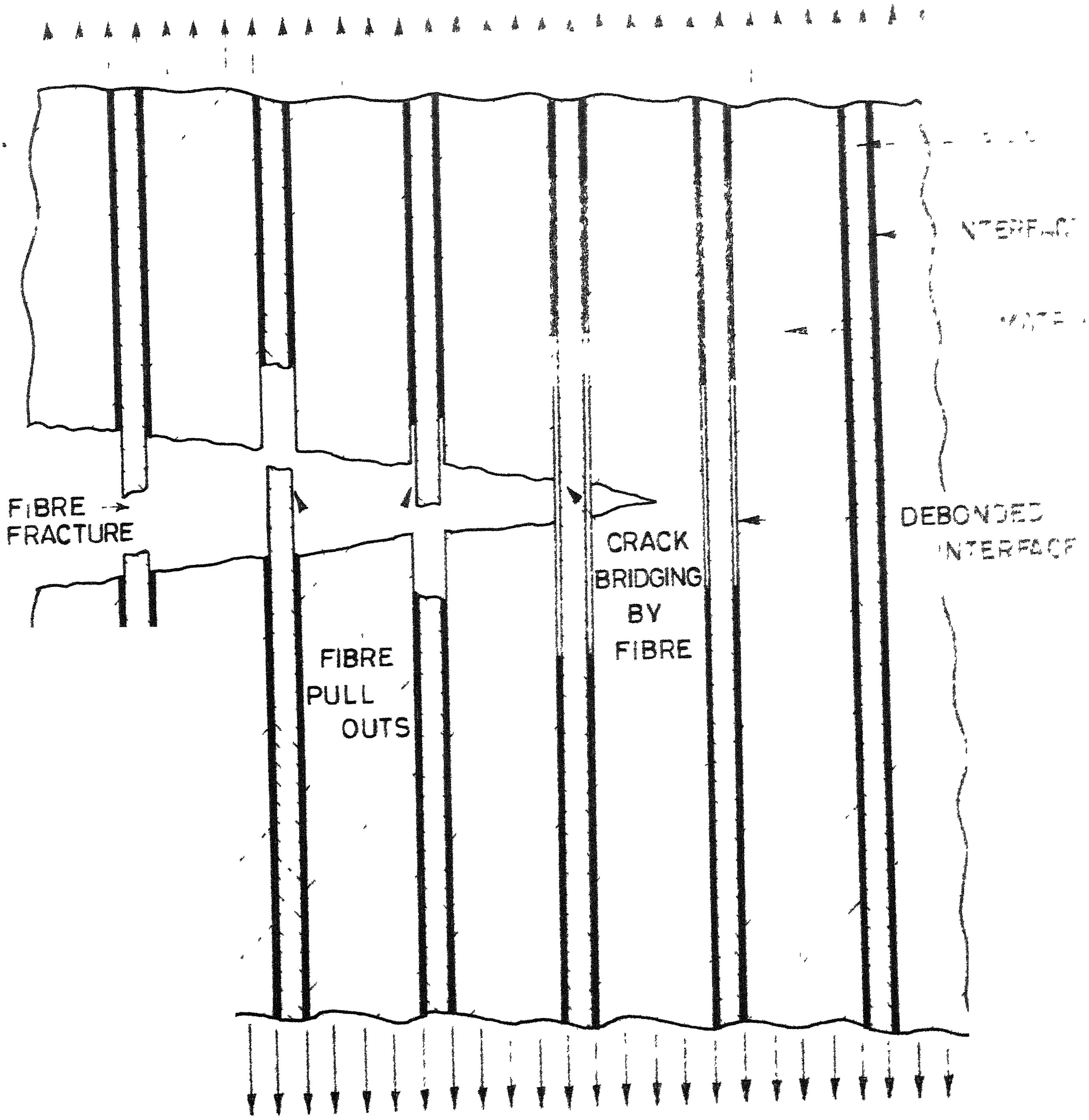
The discipline of fracture mechanics has helped the designers to dispense with the rather elusive concept of toughness, as the critical flaw size for catastrophic failure at operating temperature and stress level was defined. This led to better utilisation of material than empirically assuming a 'factor of safety' to compute the permissible stress level from the materials' strength constants. Unfortunately the basis of classical fracture mechanics makes many simplifying assumptions like material homogeniety, isotropicity and self similar crack growth under rising load conditions which are in general not applicable to aligned fibre composites and hence a direct application of the concepts of fracture mechanics which encompasses most conventional materials cannot be made in case of fibre reinforced composites.

Since the application potential of these materials in aircraft, spacecraft and other bulk carrier structures was realised, a need to develop some basis to ensure structural integrity and reliability from the point of view of criticality of cracks was felt. Although considerable efforts have been made in this direction in the last decade, the designers still have to rely on experimental data as a basis for their design. The apparent absence of simple analytical tools to predict or at least define the toughness of fibre composites is borne out by the fact that composite materials' fracture is a combination of a complex set of events in the microscopic domain whose relative contribution to the fracture process cannot be ascertained with a high level of confidence. In case of self similar crack growth as in case of metals, the high energy at the crack tip is released by plastic deformation ahead of the tip which results in crack blunting about the plastic hinge and further propagation takes place when the material in the plastic zone becomes brittle due to strain hardening. However in case of brittle fibre-brittle matrix composites the question of plastic deformation does not arise, but a large number of microfracture events occurs ahead of the sharp notch, constituting what is known as the 'damage zone' (which has been termed analogous to the plastic zone as in case of metals).

A more complex feature of composite materials' fracture is that even on a macroscopic scale the crack propagation is not self similar and visible branching of the crack takes place invariably, which thwarts any effort to determine with reasonable accuracy the area swept out by the crack. Fibre composites, by virtue of their heterogeniety and high orthotropy pose a much complex fracture analysis problem and a complete understanding of their crack growth can be possible by consideration of the micromechanical aspects of crack extension due to multimode energy dissipative processes during their failure. Some of the obvious microfracture events such as debonding due to interface failure, fibre pullout, crack bridging by fibres, apart from fibre and matrix fracture is shown schematically in Fig. 1.

1.2 Historical Development:

In this section, some key efforts in this area are examined with a critical view point. Composite fracture research falls broadly into two categories (i) A continuum analysis for a homogeneous, anisotropic, linear elastic



MATIC REPRESENTATION OF THE POSSIBLE
FRACTURE EVENTS DURING CRACK PROPA-
ON IN A UNIDIRECTIONAL COMPOSITE
A.

material containing an internal or external flaw of known length. This approach completely ignores the heterogeneous nature of the composite and the basic way the heterogeniety affects crack growth. (ii) A semiempirical analysis of the micromechanical details of the crack tip region in a simple model composite and thereby estimating the individual contributions of the various energy dissipative processes operatine on crack extension. Whereas the first approach is directed towards providing the designer with some simple physically visualizable material constants for designing for fail safe operation of composite structures, the second approach paves the way towards development of tougher and better composites to meet the stringent design requirements of advanced structures.

Wu[1] postulated that under certain conditions the techniques of isotropic fracture mechanics can be applied to unidirectional composites however, it is realised that such composites are of little use in practice. Waddoups et.al[2] attempted to apply Irwin's development of the equivalence of the energy dissipation rate and the stress intensity at the crack tip, by modelling the region in the immediate vicinity of a circular cutout, an intense energy region of a particular characteristic dimension and their contention was that the normalised stress ratio was a function of the absolute dimension of the cutout and the intense energy region. However, since this theory stands on the inverse square root singularity, its general applicability in case of composite laminates is doubtful. An early critical review of the state of art of composite materials' fracture was made by Zweben[3] who basically outlined the major avenues of approach to the problem and examined their relevance and appropriateness. Sih et. al[4] studied the fracture of composites as a function of fibre volume fraction and claimed that there exists an optimum fibre concentration for which the toughness is maximum which is understandable in the sense that the material would lose its composite nature with the predominance of any one constituent. Harrison[5] tried to remove the restriction of self similar crack growth by postulating different energy release rates in directions parallel and perpendicular to the original crack and thus established the condition for axial splitting, the predominant failure mode in unidirectional laminae with axial loading.

Establishment of a fracture toughness parameter on the basis of crack growth resistance (or R-curve) approach had evoked interest of quite a few workers in this field. The experimental results of Awerbuch and Hahn[6] from unidirectional Boron/Aluminium and Borsic/Titanium were found to agree well with analytical results. However results of Gaggar and Broutman[7-12] and Agarwal and Giare[13] on random fibre reinforced plastic go contrary to each other. Whilst Gaggar and Broutman had established a fracture toughness parameter at the onset of unstable crack growth as independent of initial crack length, Agarwal and Giare have found that such a parameter varies with initial crack length and cannot be deemed to be a material constant. This approach itself should be used for materials which remain predominantly elastic with limited plastic flow before fracture and so its meet to try to investigate the fracture phenomenon for composites whose fracture is fibre controlled (little or no debonding) and should be exercised with caution while dealing with composites undergoing interfacial shear failure dominated fracture.

Two related criteria for predicting fracture under uniaxial tension, based on the normal stress distribution ahead of a flaw were proposed by Whitney and Nuismer[14] which were shown to corroborate with limited experimental . data. They noted that since unlike metals, a positive correlation does exist between the notched strength and the materials' unnotched strength i.e. more the unnotched strength , more the fracture toughness; resorting to complications arising in the extension of linear elastic fracture mechanics (LEFM) for composites could be dispensed with. The criteria known as the 'Point' and 'Average' stress failure criteria are subsequently explained and shown to agree reasonably well (taking into account the statistical nature of composite materials' fracture) with a fairly large number of different brittle fibre/brittle matrix composite laminate systems with various stress environments, including compression.[15-17] These criteria have also been substantiated by results of Agarwal and Giare [13] for straight cracks in random fibre reinforced epoxy and Daniel[18] for round holes in graphite/epoxy systems. These criteria are perhaps the most attractive for the purpose of predicting the fracture stress in presence of flaws in real structures. The simplicity of the theory and establishment of the constancy of the characteristic distance for a generalised system of laminates which has to be supported by more data, can provide the structural designer with analytical expressions for their design without resorting to concepts of LEFM. Pipes et.al [19] have introduced a new three parameter strength model for predicting the strength reduction of fibre composite laminates in presence of a circular hole and their experimental data seem to corroborate well with the model. The influence of the laminate stacking sequence on notched strength is predicted by a unique notched strength-radius superposition method, through definition of radius shift parameters, which holds good for data of others[20,2] also. While the efficacy of such semiempirical approach , where the trend is to introduce more and more disposable parameters into a continuum mechanics formulation is yet to be ascertained, a note of caution must be borne in mind before generalisations of such theories: with enough parameters any model can be made to fit a body of experimental data, but the appropriateness of such theoreis must be judged by seeing whether the parameters have any physical significance or not. If not, the model is just a curve fitting device which can be used to interpolate a given set of data. On the other hand, if the parameters have physical significance, then experiments can be set up to determine them and thus the model can be extrapolated to real life systems.

1.3 Scope of Present Efforts:

This work was aimed at judging the efficacy and applicability of the R-curve approach and the Whitney-Nuismer criteria to a type of fibre reinforcement in epoxy matrix, which has hitherto received little attention by academicians. Although woven roving mats are extensively used for reinforcement with epoxy whenever mutually perpendicular bidirectional reinforcement is required in a variety of advanced structural applications, their benefit in terms of process economy and structural efficiency has not been adequately realised in the sense that not much quantitative treatment of this type of system is reported in literature. As distinct from the conventional crossplied laminates in which distinct laminae of unidirectional lamina are bonded together by suitable adhesive, the balanced orthotropic laminate considered here has an edge because the whole laminate can be moulded as a whole, rendering any consideration of interply delamination superfluous. Since the strands of the reinforcement are interwoven, the failure mechanism under rising load is expected to be unique and as a result the fracture behaviour which depends on microscopic aspects of failure should be different than its conventional crossplied counterpart though macroscopically their unnotched strengths may not be much different if the fibre concentration and other parameters are same. With these facts in mind it was decided to try to quantify the fracture of balanced orthotropic laminate constituting woven glass fibre reinforcement with epoxy matrix in light of techniques which have so far been developed.

The test programme consisted in determining and analyzing the load-crack opening displacement plots from single edge notched (SEN) and centrally situated round holed samples. All the tests were carried out in the opening mode or Mode I type of loading.

The notched strength data was obtained from samples cut from bidirectionally reinforced epoxy composite at four different orientations with respect to the loading axis with a view to study the fracture behaviour of the material as a function of fibre alignment. The tests on centrally holed specimens were made on two different grades of epoxy-one brittle and the other with a small quantity of additive, dibutyl pthalate (DBP) which is known to impart some ductility to the matrix.

CHAPTER TWO

SAMPLE FABRICATION, TESTING AND MATERIAL CHARACTERISATION

2.1 Sample Fabrication:

2.1.1 The Constituents:

The composite material plates moulded for the test programme consisted of epoxy resin as the matrix material and E-glass woven roving as the reinforcement. The reinforcement type is same throughout the test programme whereas tests are conducted on three types of matrix material namely, Araldite CY 230, LY 553 and LY 553 with the addition of 2.5% Dibutyl Pthalate. The epoxy resin was supplied by CIBA GEIGY (INDIA) and the woven roving by FIBREGLASS PILKINGTON (INDIA). The hardener used was CIBA GEIGY HY 951 and was used from one batch for fabricating all the plates.

2.1.2 Plate Moulding:

The epoxy resin and the hardener are accurately weighed out in the ratio of 10:1 respectively in clean glass beakers. The resin is then heated to about 60°C by an infrared lamp to drive out air bubbles, and then allowed to cool to room temperature. The glass fibre fabric was cut approximately 45 cm by 25 cm in size and twelve such pieces were required for each plate to achieve the desired volume fraction.

The mould consists of two heavy 25 mm thick MS plates, whose one face is given a smooth surface finish and both have three holes each on parallel sides which match when flushed face to face, through which 12.7 mm diameter bolts are inserted for clamping. Two clean Mylar sheets are then adhered to the mould faces with grease, between which the composite is cast facilitating easy stripping and providing good surface finish of the cured material.

The hardener is slowly added to the resin and stirred with a stainless steel spatula without turbulence to avoid frothing accompanied with bubble entrapment. The stirring is continued till uniform mixing is achieved and no streaks are tobe seen in the liquid. A small quantity of resin is poured into the bottom mould plate and a piece of the cut reinforcement mat is placed on it and the resin impregnated by help of a rubber roller. In this manner the resin and the layers of the glass fabric are laid on top of one another successively till all the twelve layers are properly wetted. Care was taken to ensure that the fibres are parallel in successive layers so as to make the laminate have two well defined principal directions. After this the top Mylar sheet is placed on the cast material and the entrapped air and excess resin are driven out by pressure by squeezing through a rubber roller run over it. The top mould plate is then clamped onto the bottom by tightening the nuts meant for the purpose. The thickness of the plate is controlled and its uniformity ensured by keeping brass spacers of 3.3 mm thickness around the bolts, between the two mould plates. The material is allowed to set for twenty four hours before being taken out and left to cure at room temperature for ten days.

Ten plates of composites using resin CY 230 and three plates each with LY 553 and LY 553 + 2.5% DBP were cast to provide the samples for the test programme. A few plates of the resins alone were cast in closed moulds for the purpose of their characterisation.

2.1.3 Sample Preparation:

For the tests on the notched samples, the specimens were cut at four different orientations, namely 0°, 15°, 30° and 45° to the pricipal fibre direction. The specimens were cut from the plate by diamend impregnated, water cooled wheel. The notch of various depths were made on the samples with a 0.2 mm thick high speed cutter. The 210 mm long, 25.4 mm wide, single edge notched (SEN) specimen used is shown in Fig. 2(a).

For tests on centrally holed specimens, the samples were cut at 0° to a principal fibre direction only. The holes of different radii were made first by drilling with a high speed tool and then enlarged to the required size by a stone cutter on a routing machine, so as to ensure that no damage occurs to the material on the exit side of the drill bit. Two small aluminium knife edge pieces were joined onto the specimen surface with adhesive cement for fixing the clip gage extensometer to measure the specimen strain about the hole diameter along the load axis. These specimens were 210 mm long and 25 mm wide as shown in Fig. 2(b).

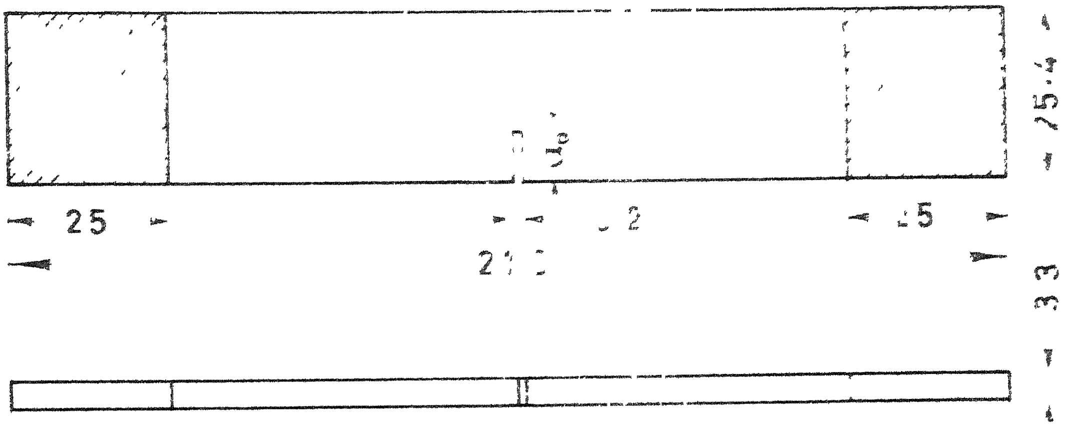
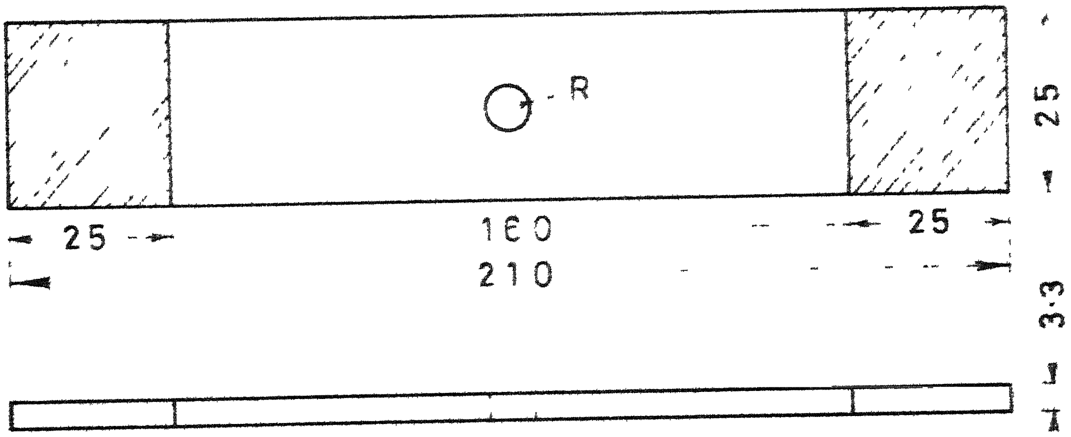


FIG 2(a) SINGLE EDGE NOTCHED SPECIMEN



ENTRALLY HOLED SPECIMEN.

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TRY AND DIMENSIONS OF SPE-
S USED IN THE TEST PROGRAMME.

The tension coupons to determine material properties including unnotched tensile strength were essentially straight sided specimens of the same dimensions fixed with square tabs at the ends to prevent failure at the grips.

The average from five identical samples were taken to establish one data point and hence the results described in this report are derived from tests on a total of about 170 samples.

2.2 The Testing System:

All the fracture toughness tests were performed on a MTS closed loop servohydraulic system. To avoid any dynamic loading effects, the tests were conducted at a constant loading rate of 32.7 Newtons per second (200 Kgs. per minute). The crack opening displacement (COD) was measured by an MTS axial extensometer (gage length = 25 mm) on the side of the specimen symmetrically across the crack mouth. The displacement across the hole diameter, parallel to the loading axis was measured by an MTS clip gage extensometer (gage length = 5 mm) on the aluminium knife edge tabs fixed for the purpose. The plot of load vs. COD for each sample was obtained on a X-Y plotter while the fracture load recovered from the system memory after each test. The specimens were hydraulically gripped over a length of 25 mm at each end using a grip pressure of 400 psi. All the tests were performed at room temperature. Photograph showing the MTS load frame and control console is shown in Fig. 3.

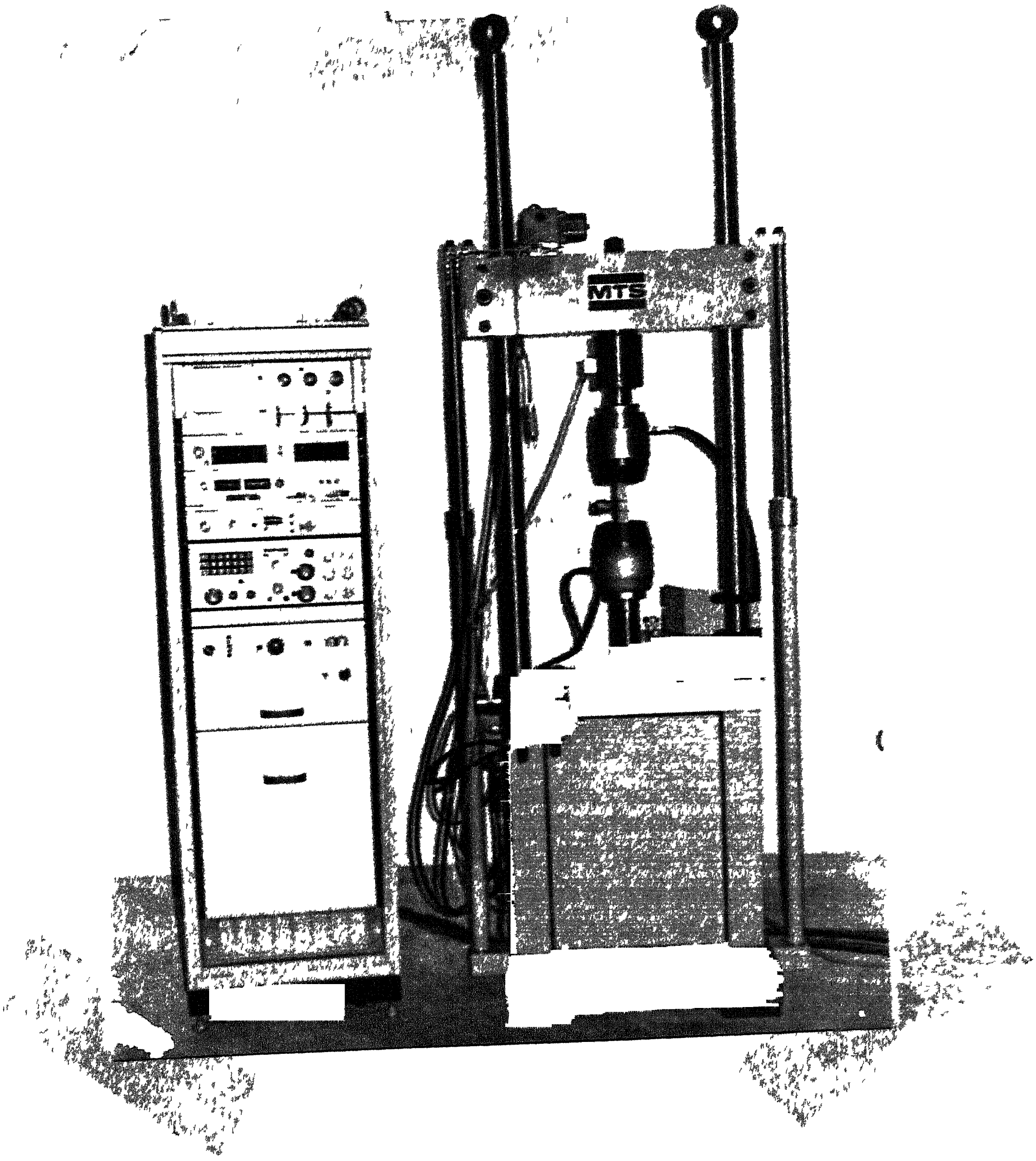


Fig. 3. Photograph showing the 100 KN MTS load frame and control console.

2.3 Material Characterisation

The volume fraction of the constituents of the composites having different matrix materials was determined by the 'burn out test', the results of which are given in Table 1 below

Table 1: Volume fraction of the constituents

Matrix material	Volume fraction (%)		
	Fibres(V_f)	Matrix (V_m)	Voids (V_v)
CY 230	58.01	41.32	0.672
LY 553	58.21	40.55	1.236
LY 553+ 2.5% DBP	58.11	40.88	1.008

As noted in Table 1 above, although the fibre volume fraction remains almost the same for different composites, the matrix and the void content change according to the nature of the matrix. Since the CY 230 resin has the least viscosity, the impregnation is best and we have the maximum content of the matrix and minimum voids. LY 553 is quite a viscous resin and hence the void content is high, but with modification with 2.5% DBP which thins down the resin, the void content is slightly reduced, but not as low as that with CY 230.

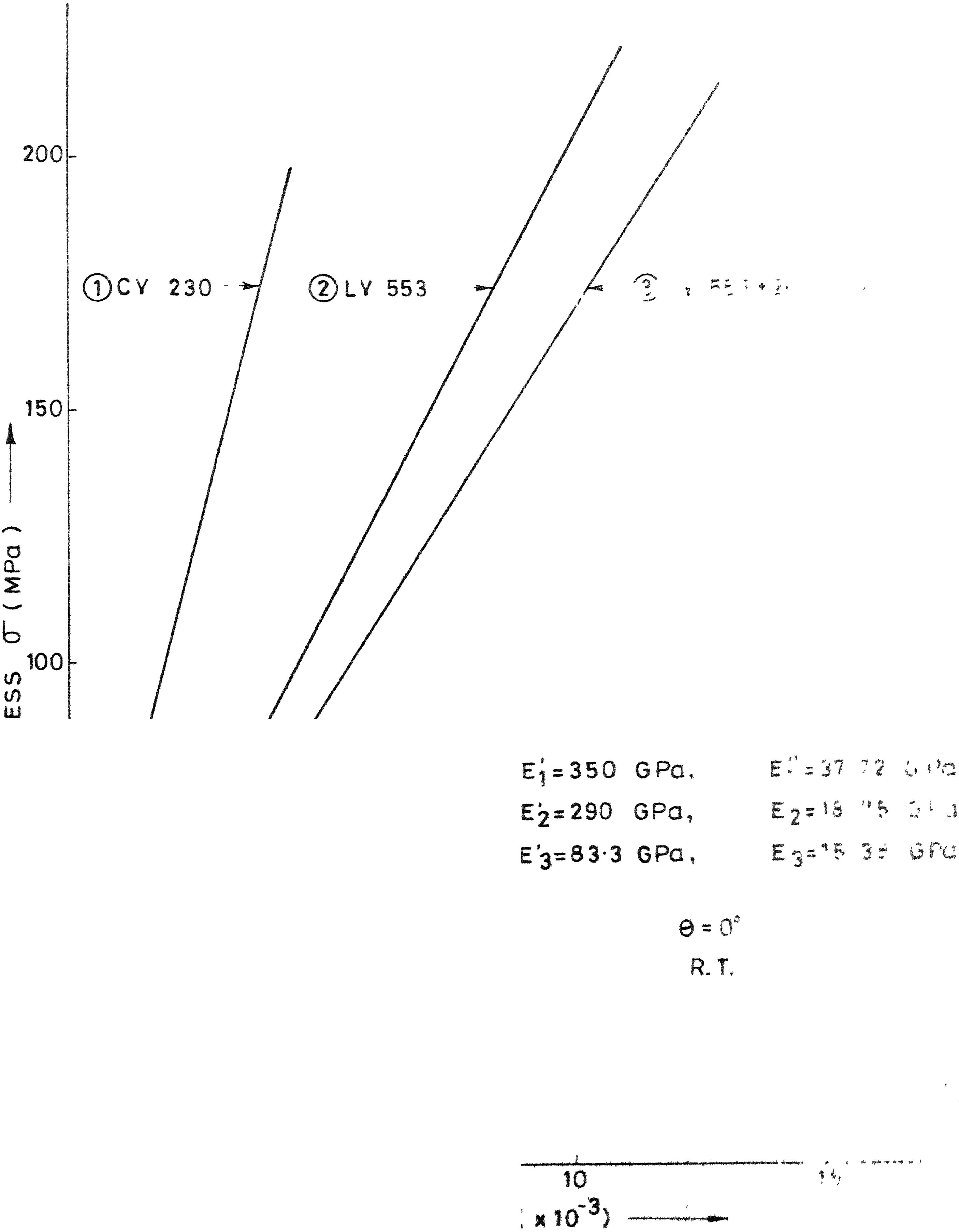
E_L and ν_{LT} were determined from a tensile test on specimens with fibres parallel to the loading axis. The lateral and longitudinal strains(for the purpose of determining the Poisson's ratio) were determined from the output of electrical resistance strain gages on a Vishay strain Indicator in addition to which the stress vs. longitudinal strain was measured by the MTS load cell and axial extensometer. G_{LT} was calculated indirectly by first determining off axis modulus in a tensile test on a 15° tension coupon and using the following relation.from Ref.21.

$$\frac{1}{E_x} = \frac{\cos^4\theta}{E_L} + \frac{\sin^4\theta}{E_T} + \frac{1}{4}\left(\frac{1}{G_{LT}} - \frac{2\nu_{LT}}{E_L}\right)\sin^2 2\theta \qquad (2.1)$$

Figures 4 and 5 give the engineering stress strain relations of the various composite materials used in the test programme. Table 2 gives the engineering constants and strength for the composite having LY 553 matrix and its modification with DBP.

Table 2: Engineering constants for composite having matrix LY 553 and LY 553 + 2.5% DBP

MATRIX	E_L' (GPa)	E_L'' (GPa)	G_{LT} (GPa)	ν_{LT}	$(\epsilon_{cu}) \times 10^{-3}$	σ_{cu} (MPa)
LY 553	290	18.75	1.376	0.21	11	222.04
LY 553+ 2.5%DBP	83.3	15.38	1.218	0.19	13	215.09



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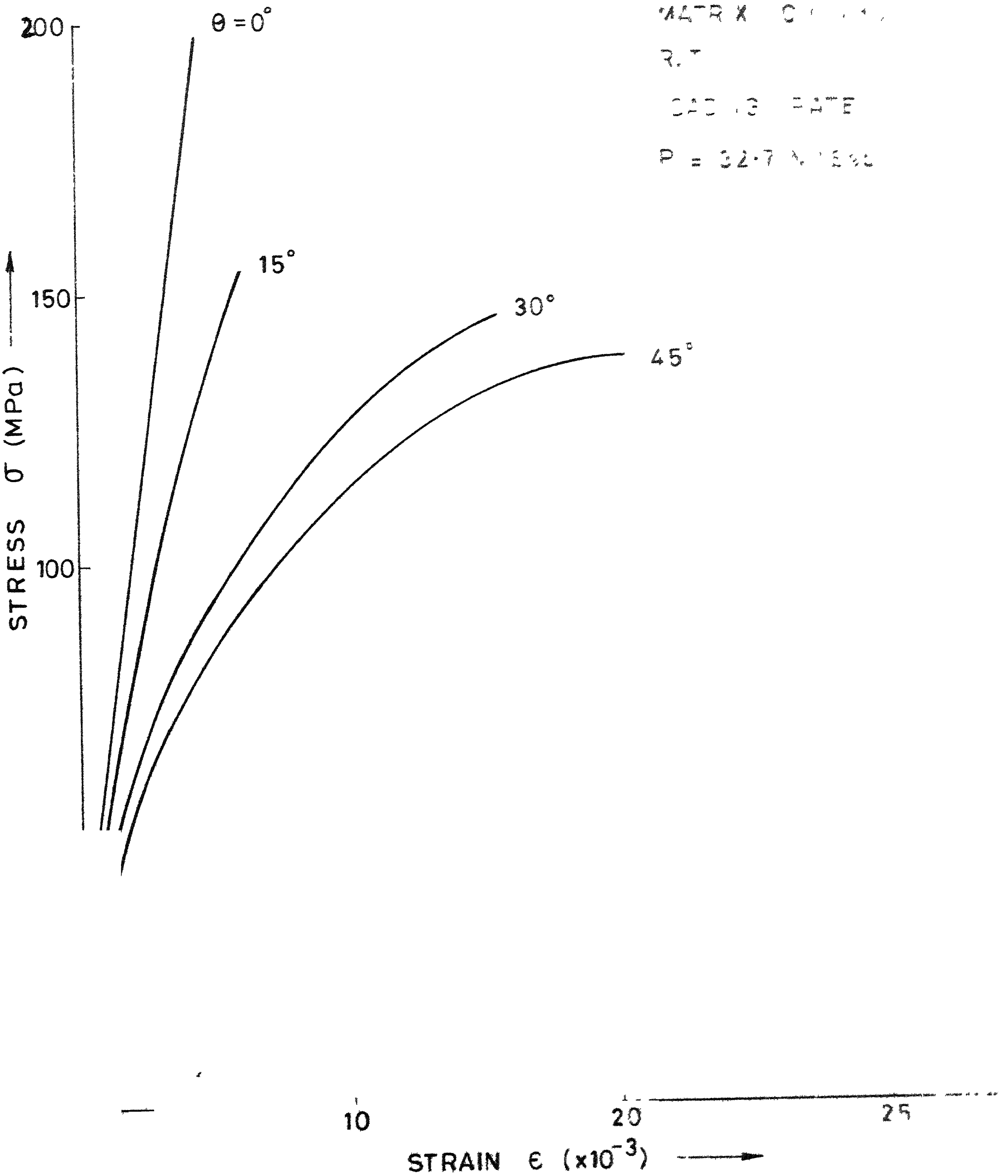
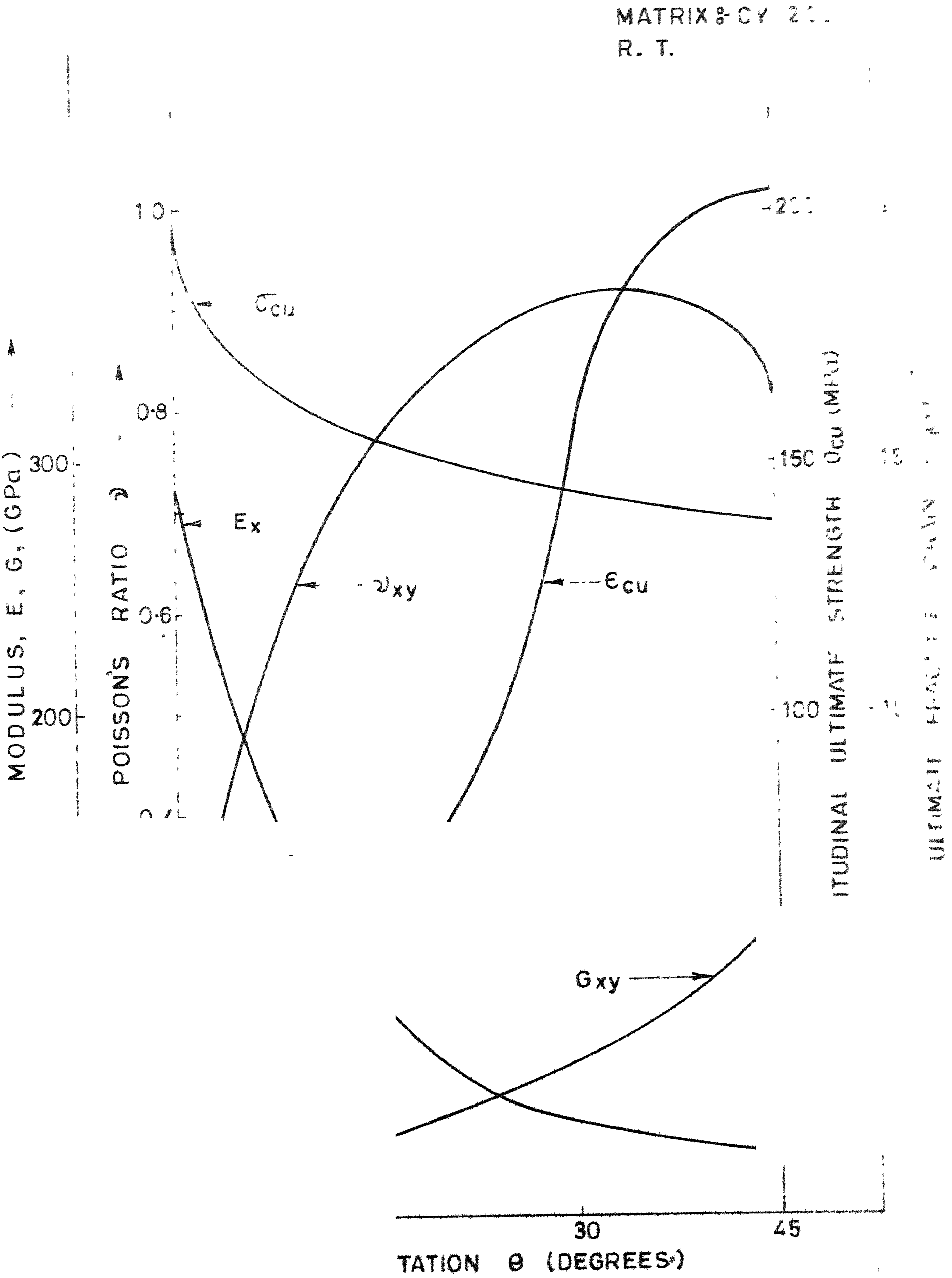


G. 5. STRESS STRAIN CURVES FOR COMPOSITES AT DIFFERENT FIBRE ORIENTATIONS TO THE LOADING AXIS.

Since the fracture behaviour of the balanced orthotropic laminate with CY 230 was studied, the various strength and elastic constants which are determined directly or indirectly from experiments are shown in Fig. 6 as a function of the orientation of the fibres to the loading axis so that the stiffness matrix at any orientation can be readily obtained. E_x , σ_{cu} and ϵ_{cu} are all determined directly from data from tensile coupons of various orientations whereas ν_{xy} and G_{xy} were determined through the transformation equations for engineering constants of composites given in Ref. 21.



ELASTIC AND STRENGTH

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CHAPTER THREE

RESULTS AND DISCUSSIONS

3.1 Material and Test Variables:

Fracture tests were performed on composite samples containing stress concentrations in the form of straight machined cracks and holes in which the reinforcement type was kept the same. The samples containing cracks were made using the matrix, Araldite CY 230 and were tested at four different fibre orientations to the loading axis namely 0°, 15°, 30° and 45°. For each orientations initial crack lengths were varied through 2 mm, 4 mm, 5 mm, 6 mm, and 8 mm.

The samples containing holes were tested in one orientation namely 0° to the loading axis only, but two different matrix materials were used, namely Araldite LY 553 and LY 553 modified with 2.5% DBP. For each matrix material four different hole diameters namely 4 mm, 6 mm, 8 mm and 10 mm were tested.

3.2 The Analysis of Data:

The raw data comprising of load-COD plots from each sample was analysed using basically two approaches. The first one is a purely fracture mechanics approach known as the Resistance Curve method by which it was attempted to establish a material constant which can predict the onset of catastropic fracture in the notched samples.

The second approach is based on the two parameter stress failure criteria proposed by Whitney and Nuismer which predicts fracture of a composite laminate containing stress concentrations without resorting to concepts of linear elastic fracture mechanics.

3.2.1 The R-Curve Approach:

The mean of the results from identical samples were used to plot the load vs. COD for each initial crack length and orientation. The load-COD curves are given in Figs 7-10 for different fibre orientations. It is observed that for the 0° orientation the load vs. COD curves remain linear and deviate from linearity near the fracture load, whereas for other orientations the nonlinearity sets in much earlier and the degree of deviation from linearity increases as the orientation increases from 15° to 45°, which is clearly in accordance with the stress-strain relationships of these composites as depicted in Fig. 5. The nature of the load-COD plots depend on the operating microfracture mechanism ahead of the crack tip under increasing load.

For the purpose of analysis the damage zone propagation ahead of the crack tip is modelled as a self similar crack growth through a compliance matching technique as suggested by Gaggar and Broutman[7-12] and Agarwal and Giare[13] . The effective crack length estimation curves for each orientation are computed from the initial portions of the load-COD plots and is shown in Fig. 11. The compliance curves for 15°,30° and

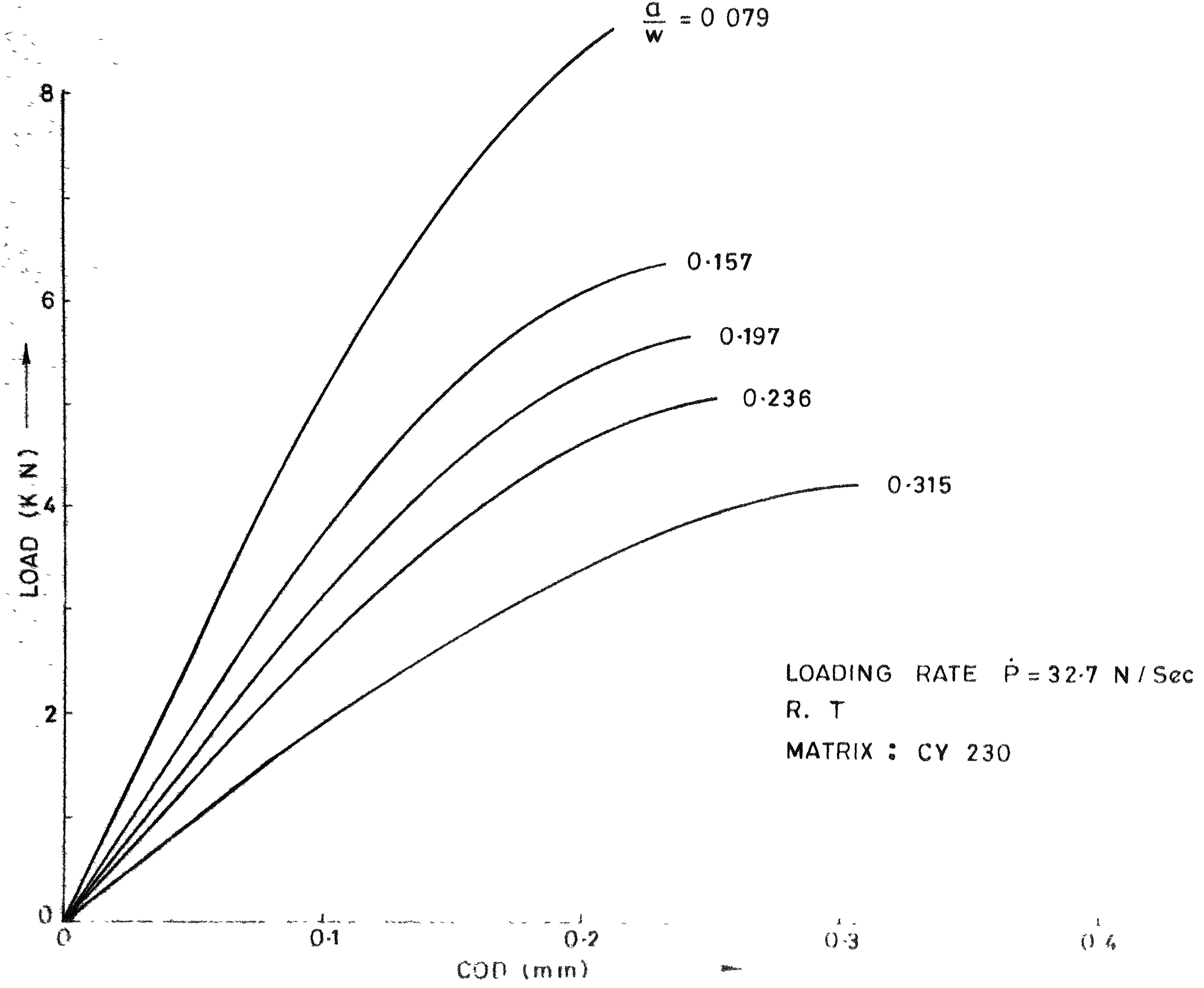
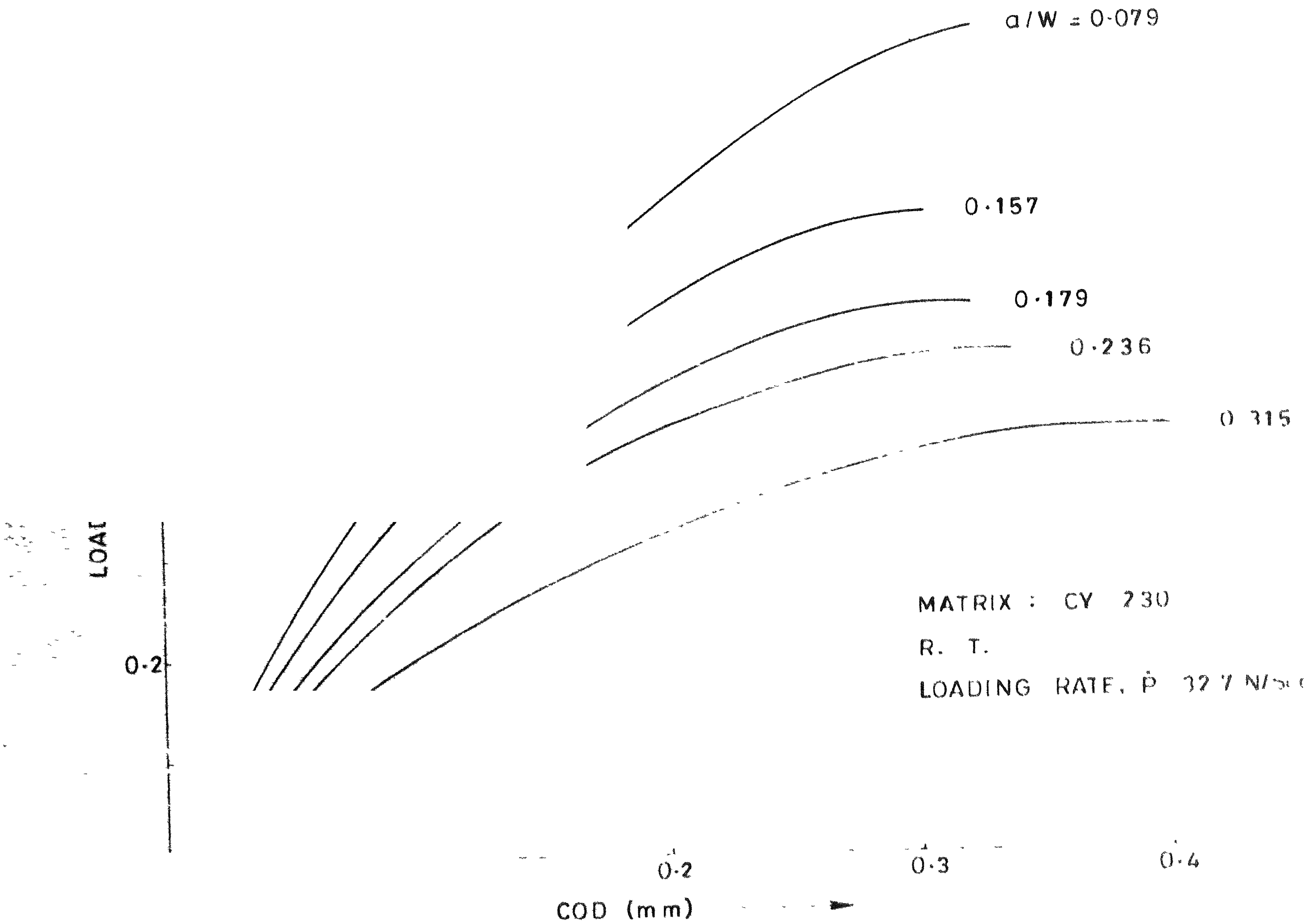
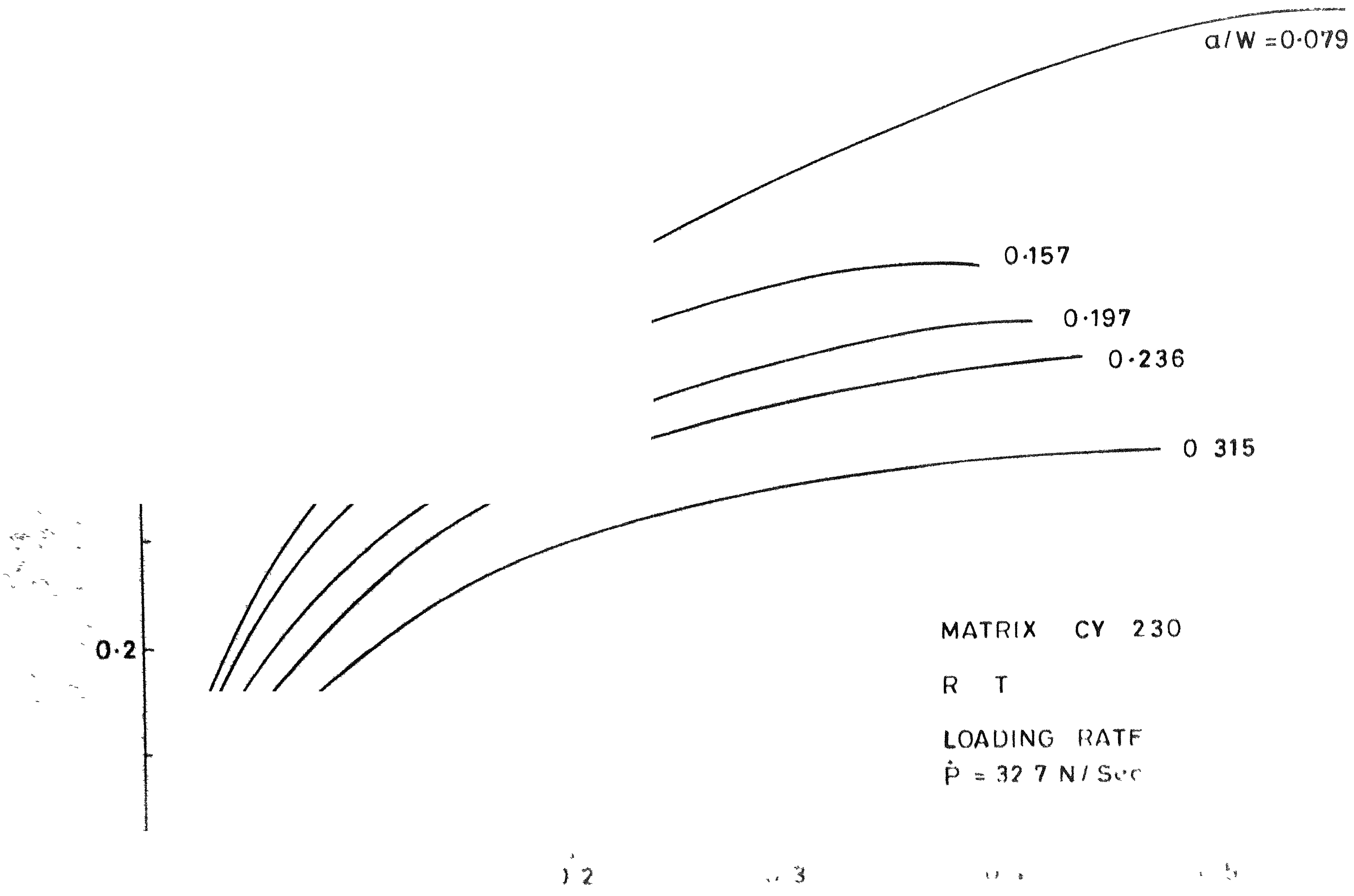


FIG. 7 LOAD VS COD FOR THE COMPOSITE WITH 0 - 0°



FOR THE COMPOSITE WITH Θ = 15°


a/W =0·079
0·157
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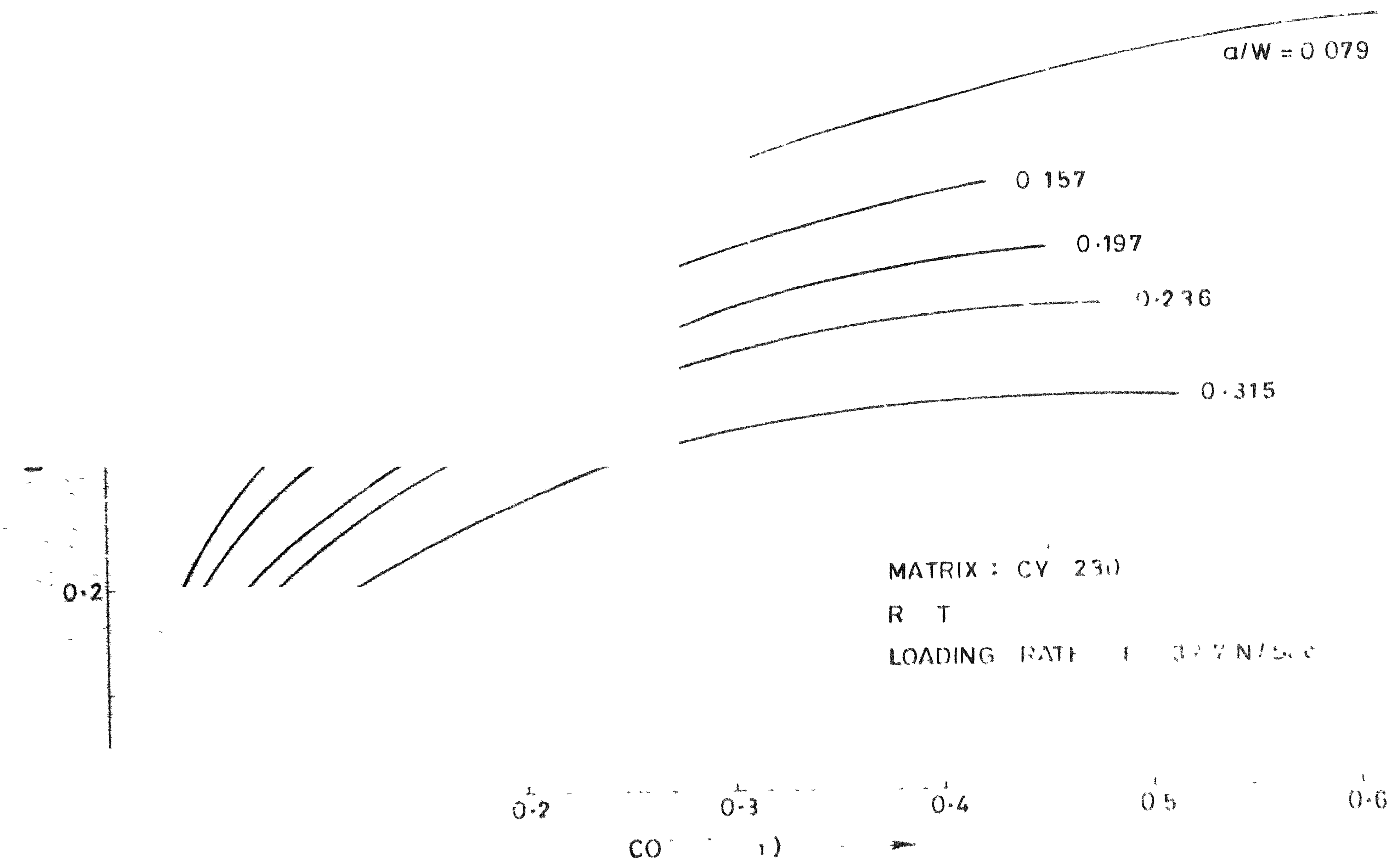
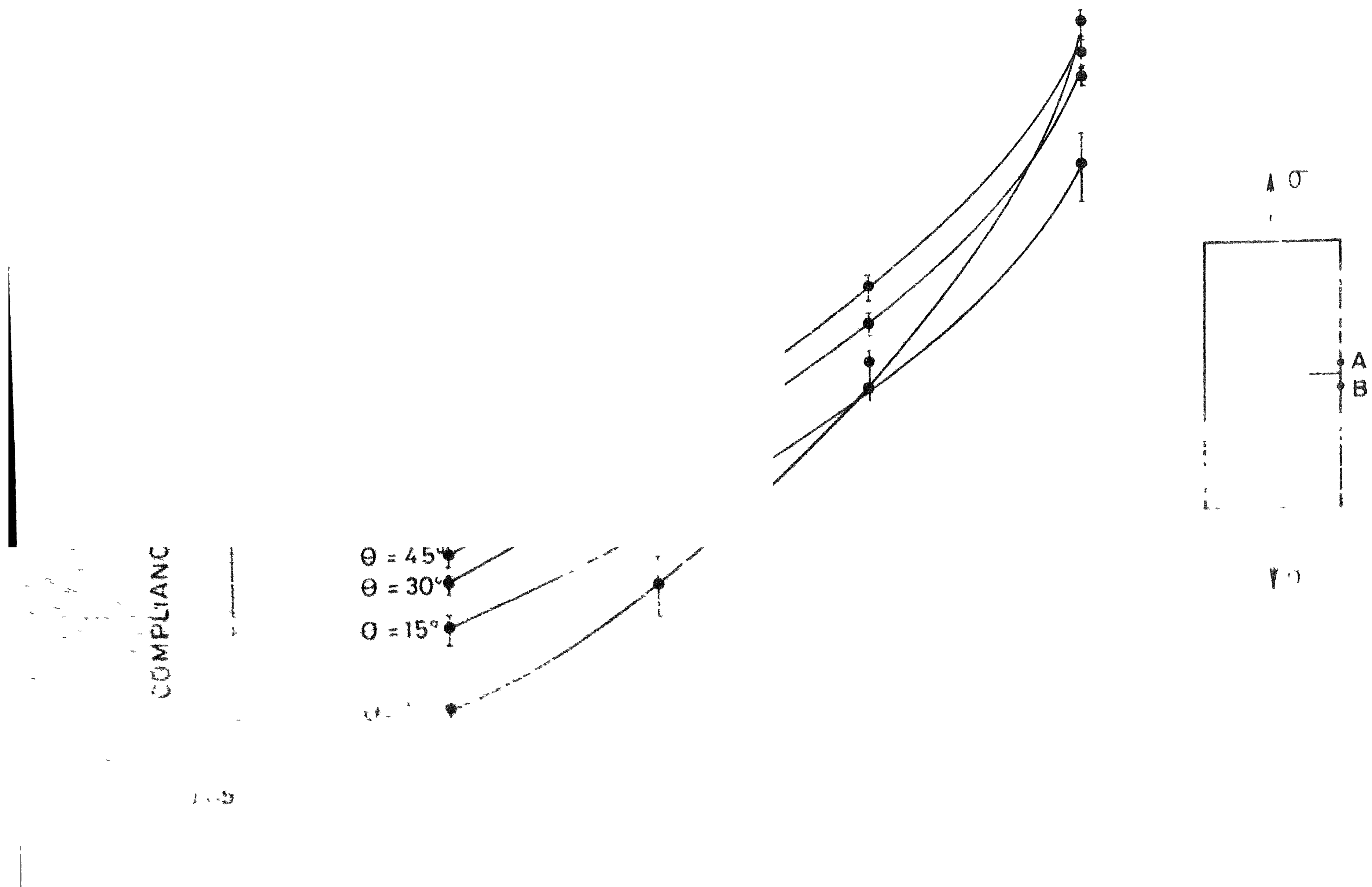
F COMPOSITE



FOR ~~MPOSITE WITH θ=45°



FIG

45^{o} are similar in nature whereas that for the 0^{o} laminate intersects the compliance curves for other orientations suggesting that the basic mechanism of fracture of the 0^{o} laminate is different from that of the off-axis ones.

The crack growth resistance curves (R-curves) for different initial crack lengths can now be constructed for each fibre orientation according to the method described by Gaggar and Broutman[9] and these are shown in Figs. 12-15. The stress intensity factor is computed from the relationship,

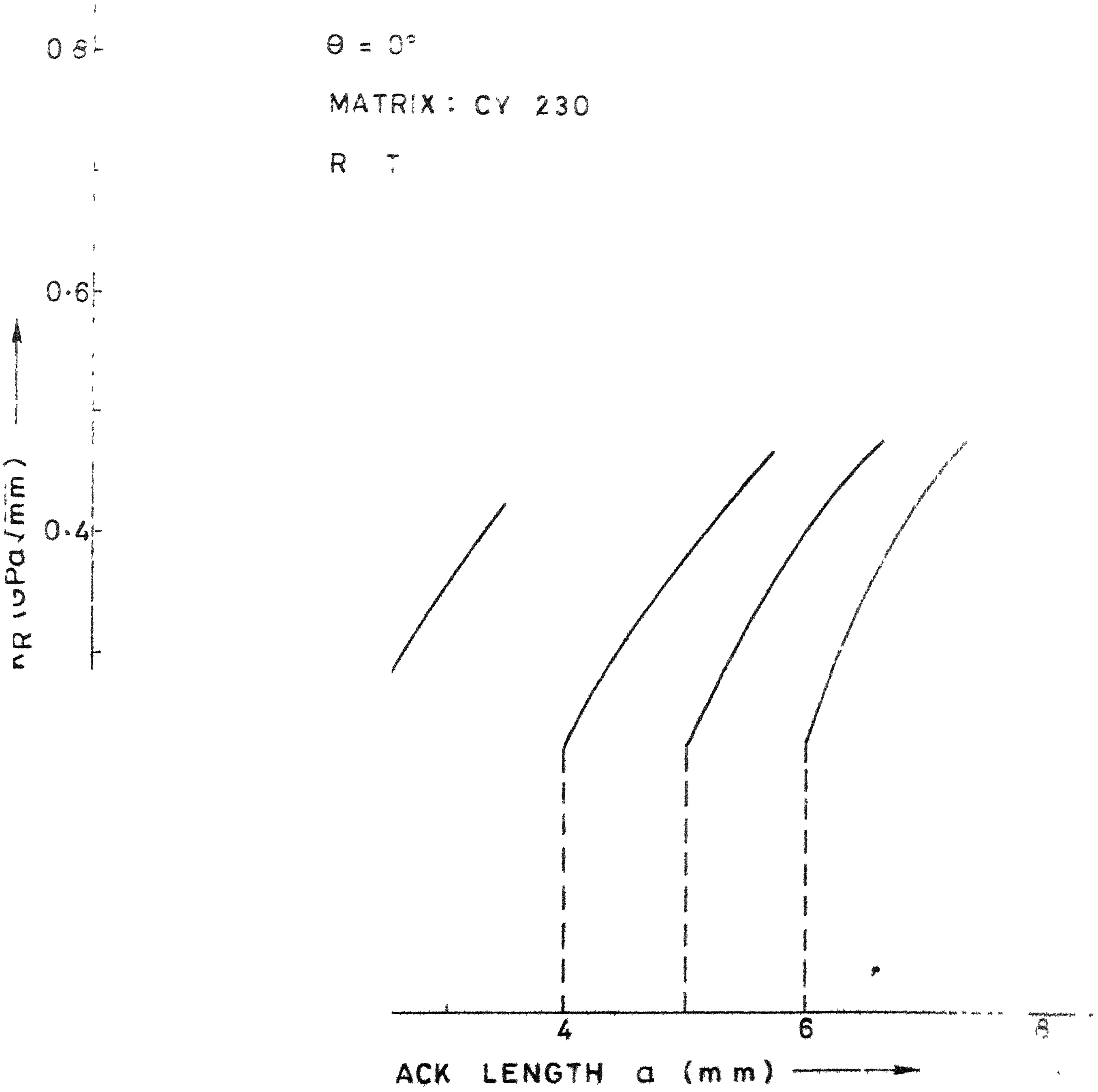
$$K_R = \frac{Pa^{\frac{1}{2}}\,Y(a/w)}{tw} \tag{3.1}$$

where 'a' is the instantaneous crack length at a particular load determined from the effective crack length estimation curve. The calibration factor, Y(a/w) for the composite is taken to be the same for isotropic materials and for SEN samples it is given by,

$$Y(a/w) = 1.99 - 0.41\,(a/w) + 18.70(a/w)^2 - 38.48(a/w)^3 + 55.84(a/w)^4 \tag{3.2}$$

Also from the Figs. 12-15 it is evident that the operating energy mechanism at the crack tip is different for the 0^{o} laminate and that of the other orientations, because the curvature of the R-curves changes sign as we go from the 0^{o} to the other orientations.

The parameter $(K_R)_{instability}$ is determined as the value of the stress intensity factor corresponding to the



GROWTH RESISTANCE CU
MPOSITE WITH θ = 0°

I.I.T. K
CENTRAL
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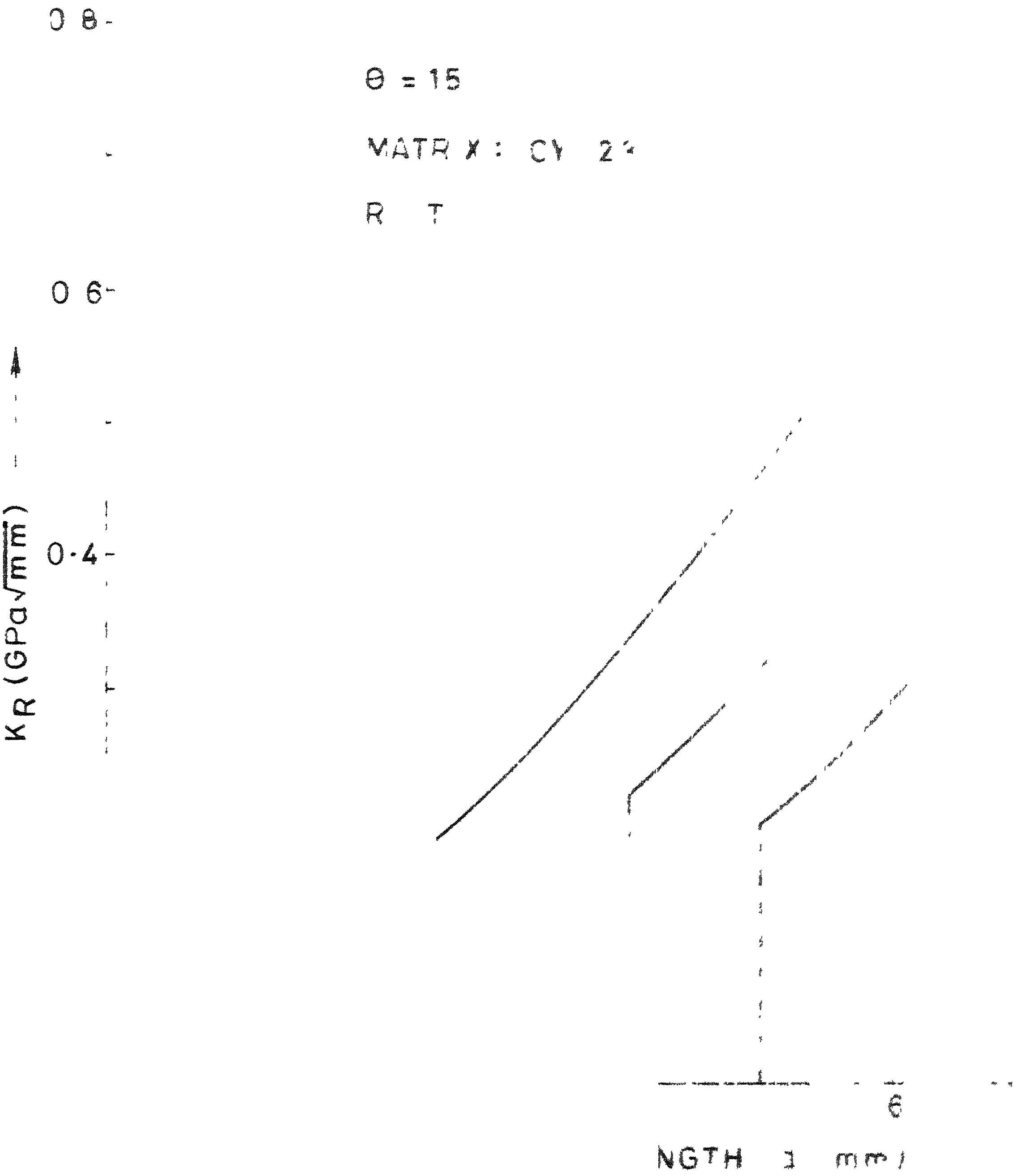
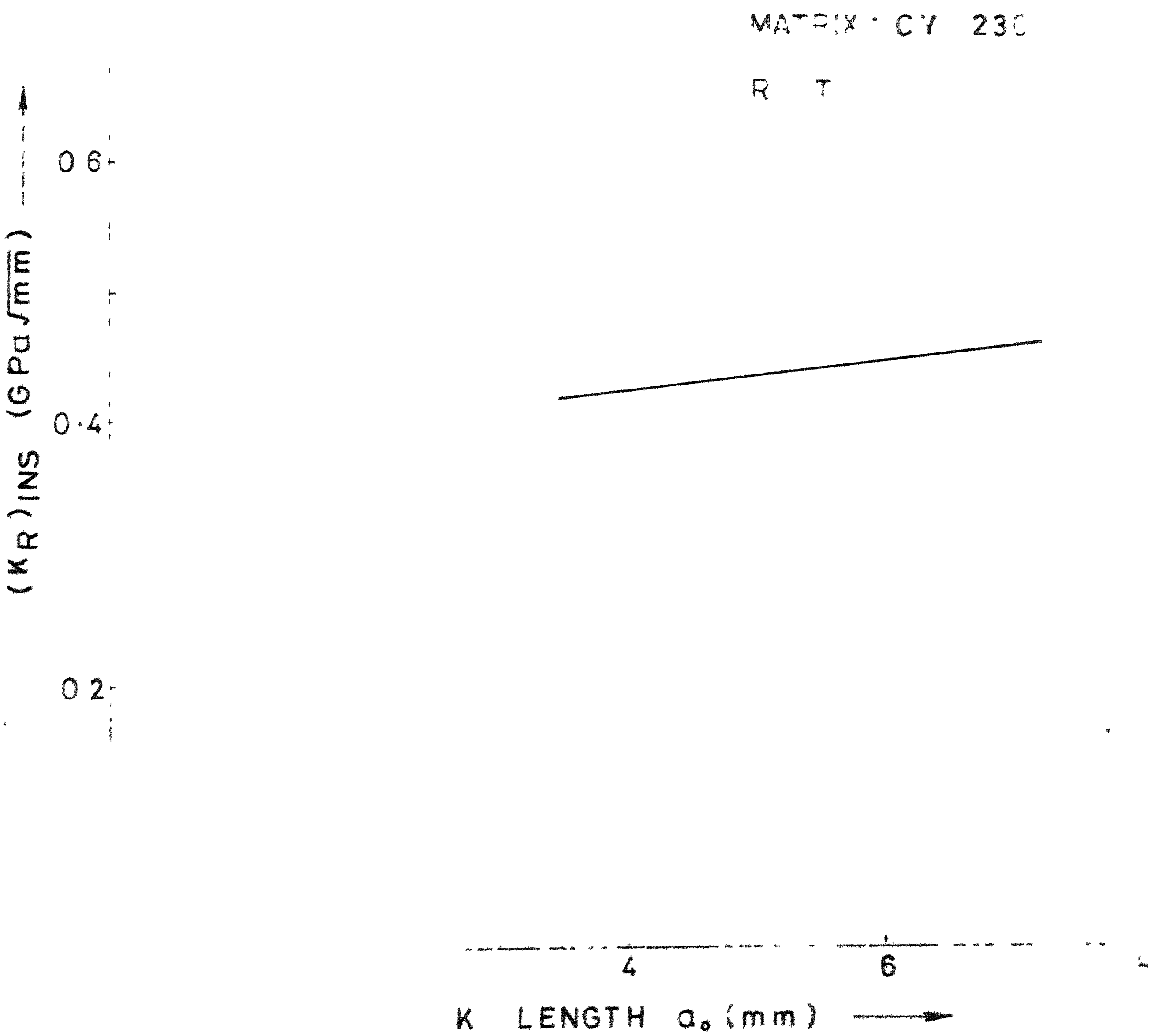


RESISTANCE CURVE [illegible]
WITH θ = 15°

point of tangency between the R-curve and crack force driving curves for constant loads near the fracture load. The point of tangency was determined by the computer and it was found that the solution existed only for the 0° laminate. The nature of the crack force driving curves and the R-curves for 15°, 30° and 45° orientations prevents attainment of tangency and so the applicability of this technique as such, to non-linear composite material systems becomes questionable. Since the R-curves are dependent on orientation it is logical to assert that the crack force driving curves must also be θ dependent, at least through the incorporation of θ in the expression for Y calibration factor which was taken for the isotropic case in this analysis. This method originally developed for linearly elastic isotropic materials like metals, can be expected to yield satisfactory results for isotropic composite systems such as short fibre composites. In the present case it is observed that the parameter $(K_R)_{instability}$ can be determined in the case of the 0° laminate only which is a predominantly linear elastic system as shown in the stress strain relationship (Fig. 5). $(K_R)_{instability}$ for 0° laminate increases slightly with the initial crack length as shown in Fig. 16. However, the increase is so small that it may represent experimental scatter and it cannot be concluded with confidence that $(K_R)_{instability}$ is a material property. These results are similar to those of Agarwal and Giare[13] for randomly oriented short fibre composites.



N OF (K_R)INSTABILITY WITH INITIAL
FOR THE 0° LAMINATE.

From the plot of crack growth resistance (K_R) vs. absolute crack extension ($a-a_o = \Delta a$) shown in Fig. 17 it can be inferred that the shape and nature of the curves are generally independent of the initial crack length, but due to scatter, any absolute conclusions must be made with caution only after a careful statistical analysis of a large amount of data.

3.2.2 Stress Failure Criteria Approach:

The stress failure criteria proposed by Whitney and Nuismer[14,15] are based on the normal stress distribution ahead of a hole or crack. The first criterion known as the 'point stress criterion' states that failure will occur when the stress at a distance, d_o, ahead of the crack tip first reaches the unnotched strength of the material, that is when,

$$\sigma_y(x,o)\Big|_{x=c+d_o} = \sigma_o' \qquad (3.3)$$

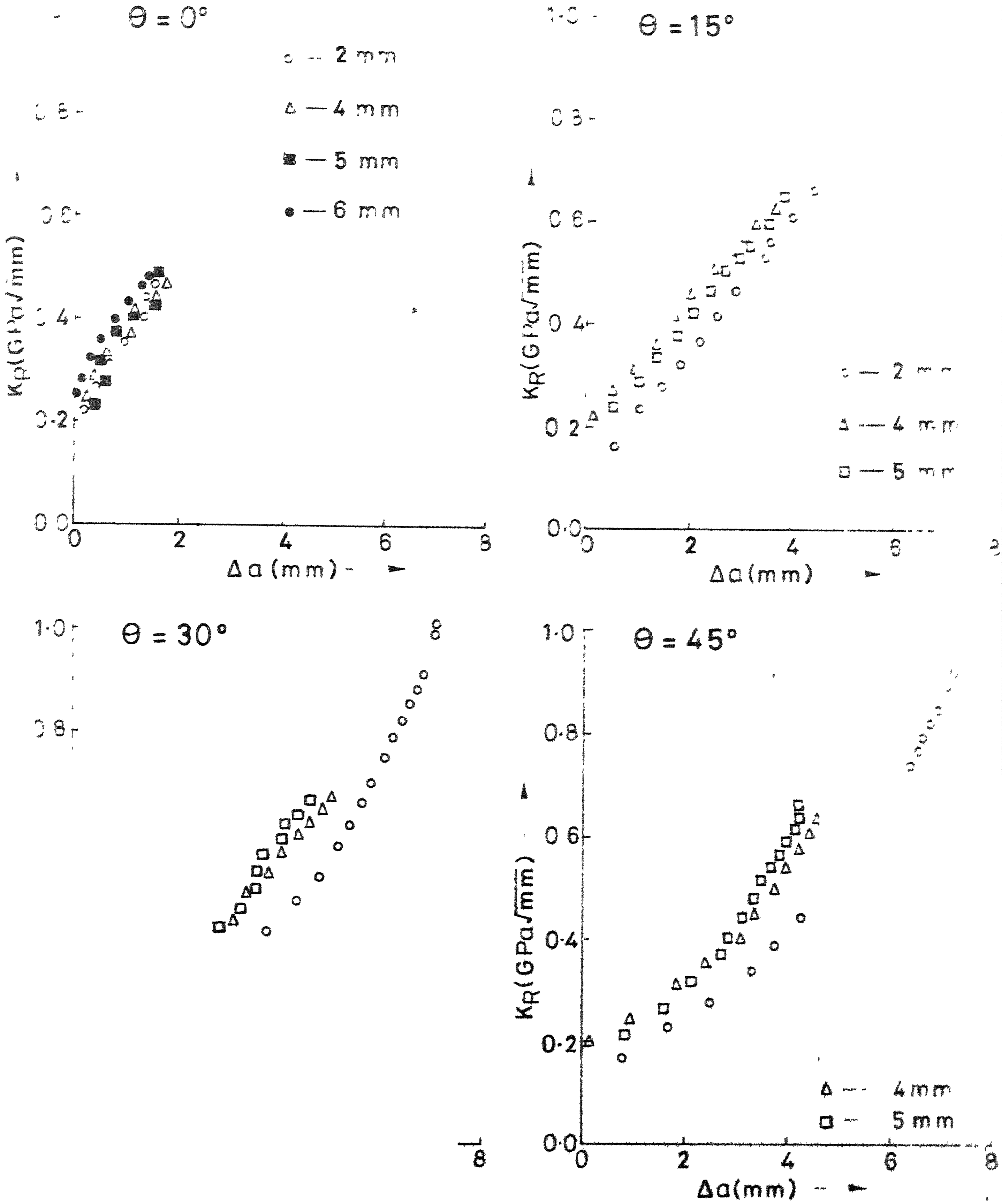
Using this criterion the ratio of notched to unnotched strength is given by,

$$\frac{\sigma_N}{\sigma_o} = \sqrt{1 - p_1^2} \qquad (3.4)$$

where

$$p_1 = \frac{c}{c+d_o} \qquad (3.5)$$

The second criterion known as the 'average stress criterion', assumes failure to occur when the average value of the normal stress ahead of the crack tip over some fixed distance, a_o, ahead of the crack first reaches the unnotched



:ESISTANCE AS A FUNCTION OF
EXTENSION. (MATRIX: CY 230, R T)

strength of the material, that is when,

$$\frac{1}{a_o}\int_{c}^{c+a_o} \sigma_y \, (x,o)dx = \sigma_o \qquad (3.6)$$

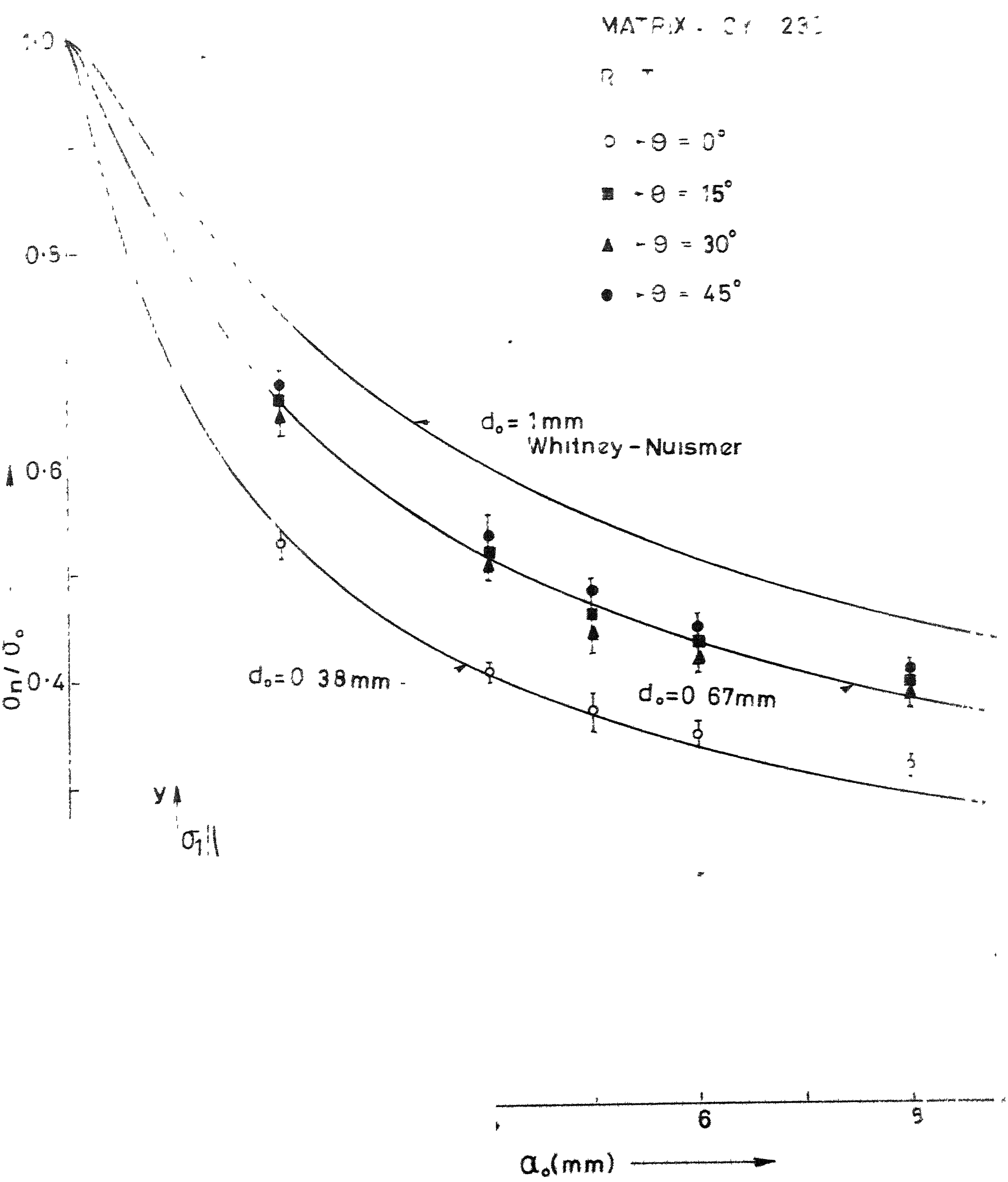
Using this criterion, the ratio of notched to the unnotched strength is given by,

$$\frac{\sigma_N}{\sigma_o} = \sqrt{\frac{1-p_2}{1+p_2}} \qquad (3.7)$$

where,

$$p_2 = \frac{c}{c+a_o} \qquad (3.8)$$

Whitney and Nuismer have tried to establish from a large data bank that the characteristic distance d_o or a_o was constant for all composite laminate systems, being equal to 1 mm and 4 mm (rounded off to the nearest millimeters) respectively. However, in the results presented here in Figs. 18-19 it is evident that although both the criteria remain valid for the balanced orthotropic system considered here, the values of d_o or a_o are smaller and distinctly different from those proposed by Whitney and Nuismer. It is also observed that the strength reduction in case of the 0^o oriented laminate is more than the off axis orientations which means that the 0^o laminate is very much notch sensative. The notched to unnotched strength ratio for the off axis orientations tend to merge in the sense that their data spread indicators overlap. This again substantiates



FAILURE CRITERION FOR

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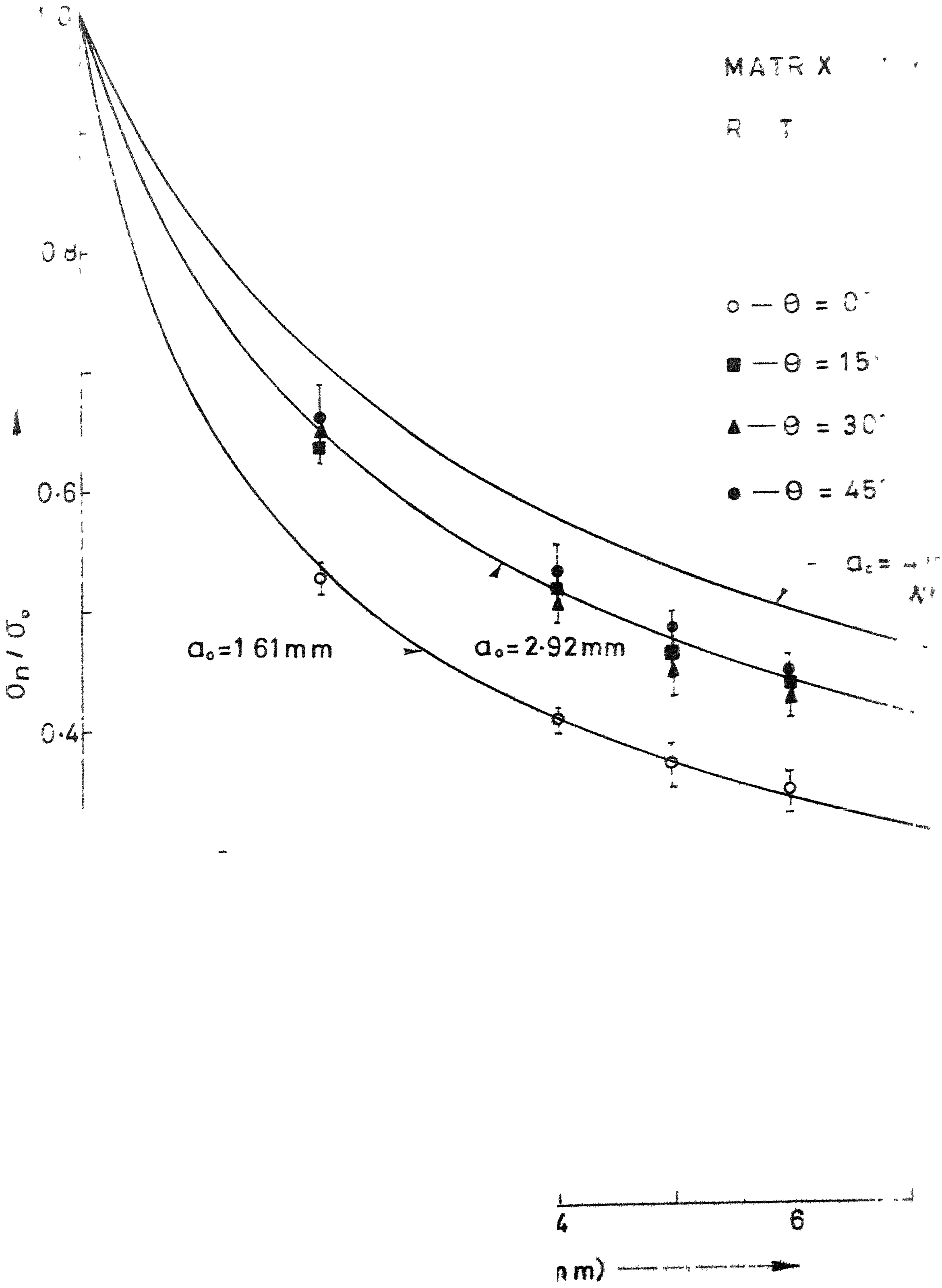


S FAILURE CRITERION

(S.

the argument that the microfracture mechanism of the off axis orientations are similar whereas those for the 0° laminate are distinctly different. The characteristic distances were determined for the 0° laminate and for all the off axis laminates combined by a least square curve fitting method using a computer programme; these are indicated in Figs.18,19. The characteristic distance may be related in some manner with the size of the damage zone ahead of the crack tip analogous to the plastic zone in case of metals. If its a measure of the damage zone size, then it can be said that the systems considered here have smaller damage zones than those considered by Whitney- Nuismer and others others.[13-18] This could be because none of them considered the type of interwoven reinforeement used here in which the damage zone is likely to remain small because the interweave of the fibres constrains the spread of damage.

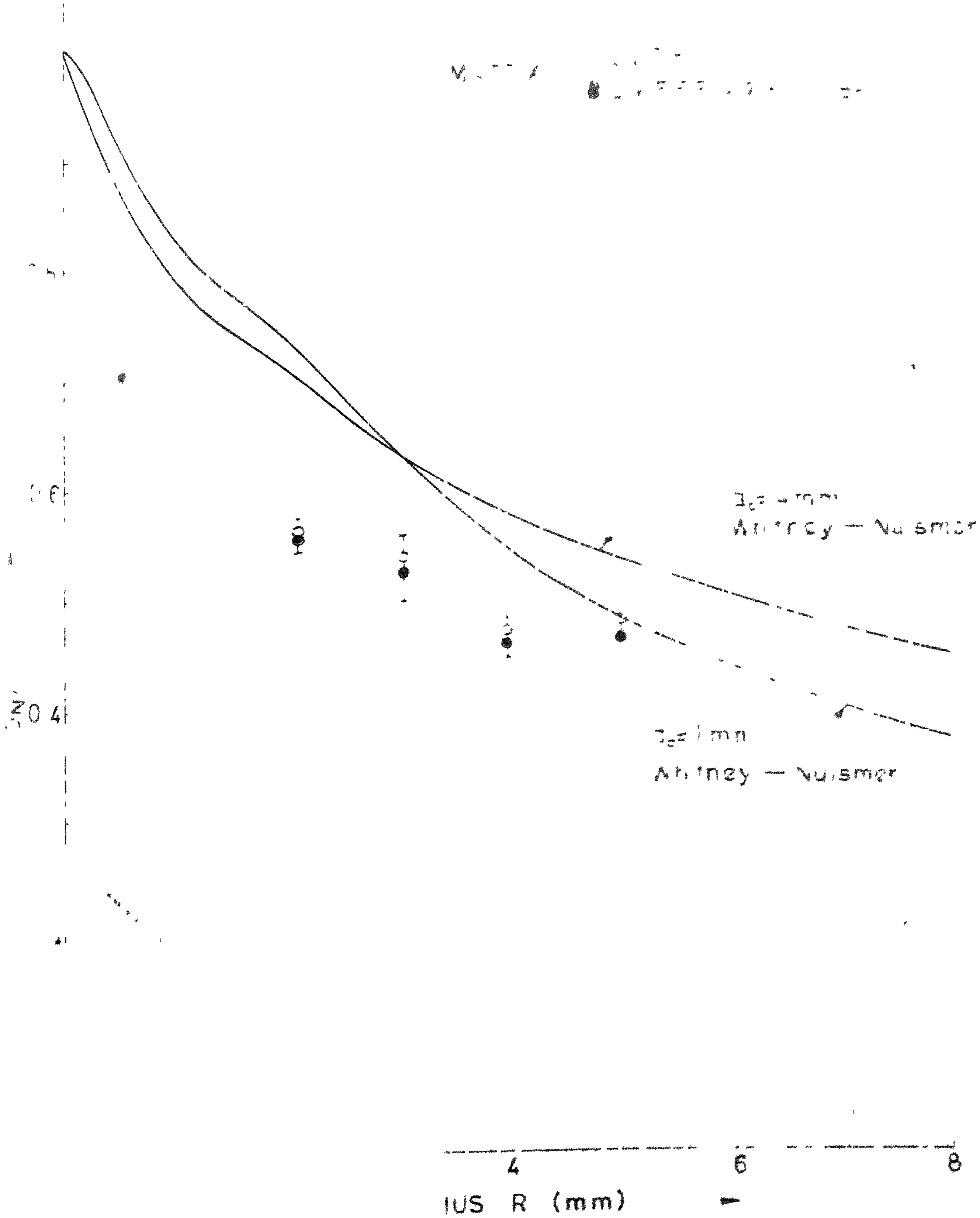
The application of the point and average stress failure criteria for holes is exactly similar in nature as for cracks, the ratio of notched to unnotched strength being given by a relation depending on the normal stress distribution ahead of a circular hole from an exact elasticity solution. It was paradoxical to note that even though circular holes of all sizes have a stress concentration factor of 3, the effect of absolute hole size on the strength was observed and this was resolved by Whitney and Nuismer by the proposition of their failure criteria.

From the results presented here in Fig. 20 it is seen that the nature of variation of the normalised strength ratio with hole radius does not corroborate with the nature of variation proposed by Whitney and Nuismer although the curves have been drawn for the characteristic distances proposed by them.

For conversion of experimental data on samples of finite width to that for an infinite plate, finite width correction factors were multiplied with the experimental normalised strength ratio as suggested by Whitney and Nuismer[15]. However Yeow and Brinson[21] noted that application of the isotropic finite width correction factors for anisotropic system could be in considerable error and that these correction factors depend on the composite material system, stacking sequence etc. This could be a possible reason for nonconformity of the results presented here with those predicted by Whitney and Nuismer.

3.3 Microfracture Events and the Path of Crack Growth:

The various microfracture events ahead of the crack tip profoundly influence the path of crack propagation as well as the macroscopic response of the material to load. This has been aptly revealed by the results presented in the previous section for the composite which is identical in all respects except the orientation of the principal fibre directions with respect to the loading axis. It is also evident from Fig. 21 which is a photograph showing fractured specimens having a



JRE CRITERION FOR HOLES

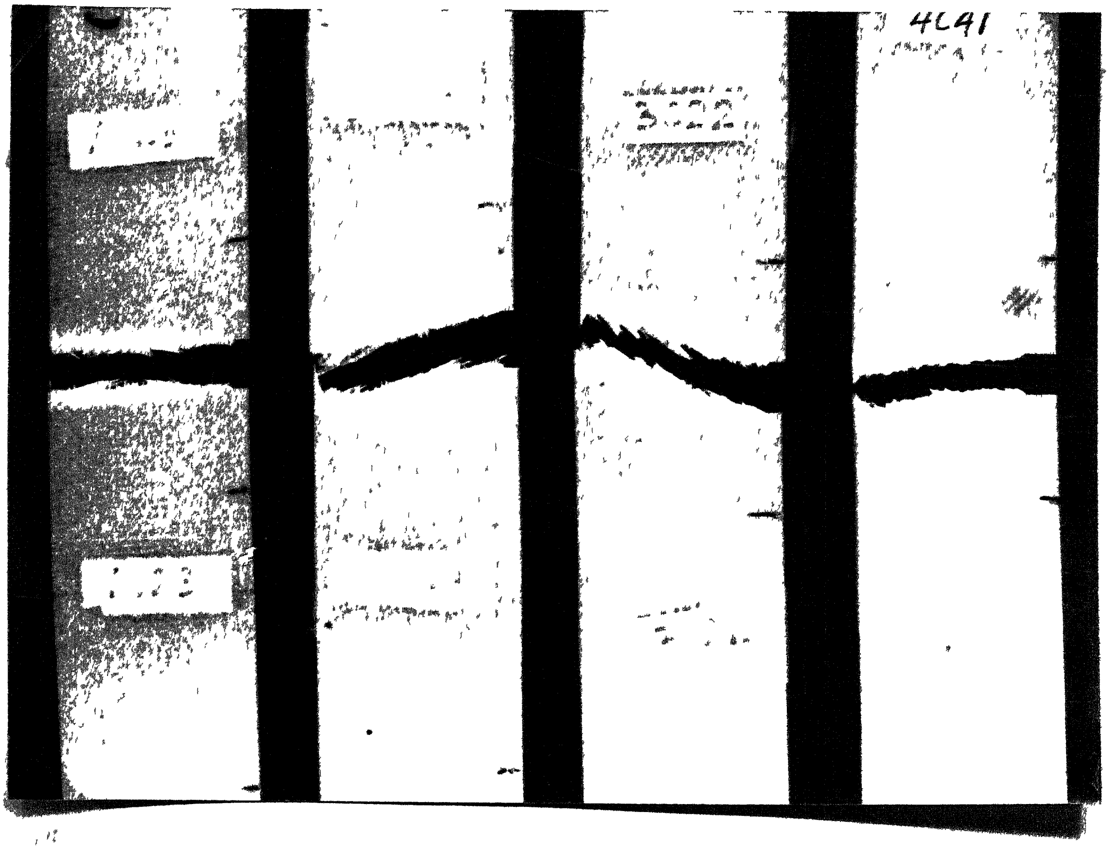


Fig. 21. Photograph showing fractured specimens of the composite with $\theta = 0^\circ, 15^\circ, 30^\circ$ and 45° (left to right) respecively.

principal fibre direction oriented at 0°, 15°, 30° and 45°(left to right) to the loading axis, that the operating microfracture mechanisms and their interactions during crack extension which is due to the particular fibre orientation influence the nature of the path traversed by the crack and consequently the resulting surface of separation. In this section an attempt will be made to rationalise the mode of crack extension behaviour in light of visual observations during the stages of crack growth during the testing process.

For the 0° laminate, where 50% of the fibres are parallel to the direction of loading and the rest perpendicular to it, the gross fracture will be controlled by tensile failure of individual fibres. The failure takes place mainly in the plane of the machined notch because of the severe stress intensity ahead of the tip and hence the susceptibility of easy debonding along the fibres perpendicular to the loading axis in the plane of the original notch. Since the fracture is fibre controlled the net elongation will be small and very little or no debonding and pullout was observed along the loading axis. This is in contrast to the fracture of unidirectionally reinforced lamina loaded parallel to the fibres, which is virtually notch insensative and separation of surfaces takes place with profuse debonding and fibre pullout to the extent that it is impossible to visualize any well defined plane of separation of surfaces. The various stages of crack growth in the 0° laminate is shown in the photographs of Fig. 22. The formation

Fig. 22. Photographs showing the stages of crack propagation in the 0° laminate.

of the damage zone is apparent by the whitening ahead of the machined notch parallel to the fibres and hence whatever toughness the material possesses is due to absorbtion of the energy at the crack tip by the formation of small microcracks parallel to the fibres.

The off-axis failure of unidirectional lamina is invariably due to interface shear and/or matrix cracking, but in the case of the balanced orthotropic laminate reinforced with woven roving mat, interaction of different failure modes occurs simultaneously. Whereas in a unidirectional lamina the surface of separation of off-axis specimens is parallel to the fibre direction, in the system considered here the crack chooses a path of least resistance which may not always coincide with a particular fibre direction. Also in the case of the conventional crossplied laminate where unidirectional laminae are stacked onto one another with their fibre orientations at 0° and 90° to each other, the off-axis failure mode will predominantly be different in different plies depending on the orientation of the fibres in that particular ply with the loading axis.

If a stress σ_0 is applied to the specimen as shown in Fig. 23 and that direction taken to be along the z axis, then the resulting stress tensor will be

$$\begin{pmatrix} 0 & 0 & 0 \\ 0 & 0 & 0 \\ 0 & 0 & \sigma_0 \end{pmatrix}$$

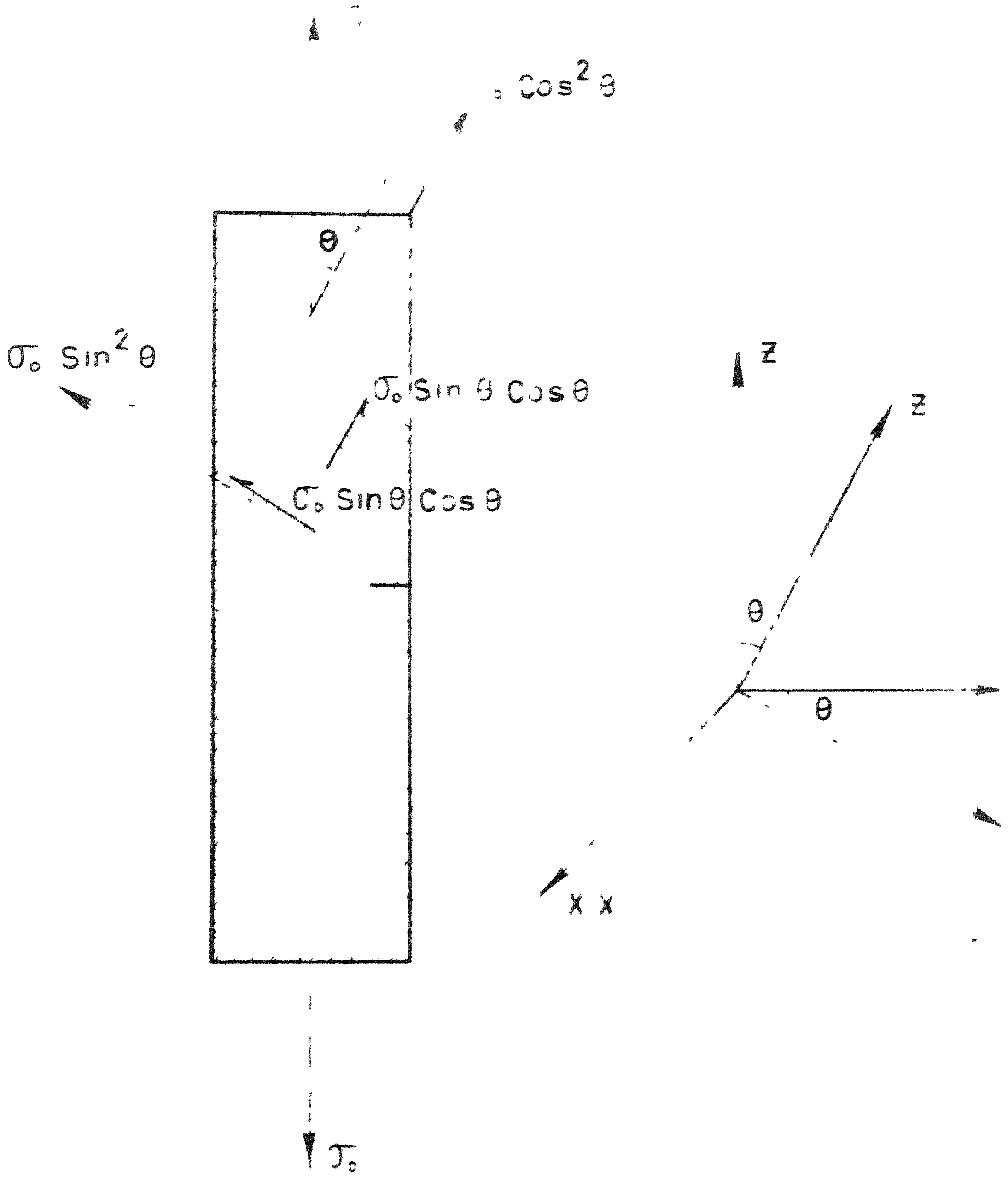


FIG 23 SCHEMATIC REPRESENTATION OF THE APPLIED NORMAL STRESS RESOLVED ALONG THE PRINCIPAL FIBRE DIRECTIONS FOR OFF AXIS SPECIMENS

To get the stresses along any direction in the y-z plane by rotation through an angle θ , the rotation matrix will be

$$\begin{pmatrix} 1 & 0 & 0 \\ 0 & \cos\theta & \sin\theta \\ 0 & -\sin\theta & \cos\theta \end{pmatrix}$$

Hence by tensor transformation the tensile and shear stresses along the principal fibre directions will be given by

$$\begin{pmatrix} 0 & 0 & 0 \\ 0 & \sigma_0\sin^2\theta & -\sigma_0\sin\theta\cos\theta \\ 0 & -\sigma_0\sin\theta\cos\theta & \sigma_0\cos^2\theta \end{pmatrix}$$

and they are shown in Fig. 23. It is observed that the shear stress along the fibre-matrix interface along both the fibre directions will remain same whatever be the value of θ. Although it is observed in the photographs in Figs 24-26, that the crack chooses to propagate along a particular interface, that it fails by shear can be ruled out, and thus there is some other factor which compels the crack to propagate along that particular interface. When θ is less than 45°, the tensile stress on the interface making an angle of θ to the machined notch is more than that on the interface making an angle of θ to the load axis. Hence it was envisaged that due to the higher tensile stress on the interface oriented at an angle of θ to the original notch it fails first and then ultimate separation of surfaces takes place by tensile failure of the

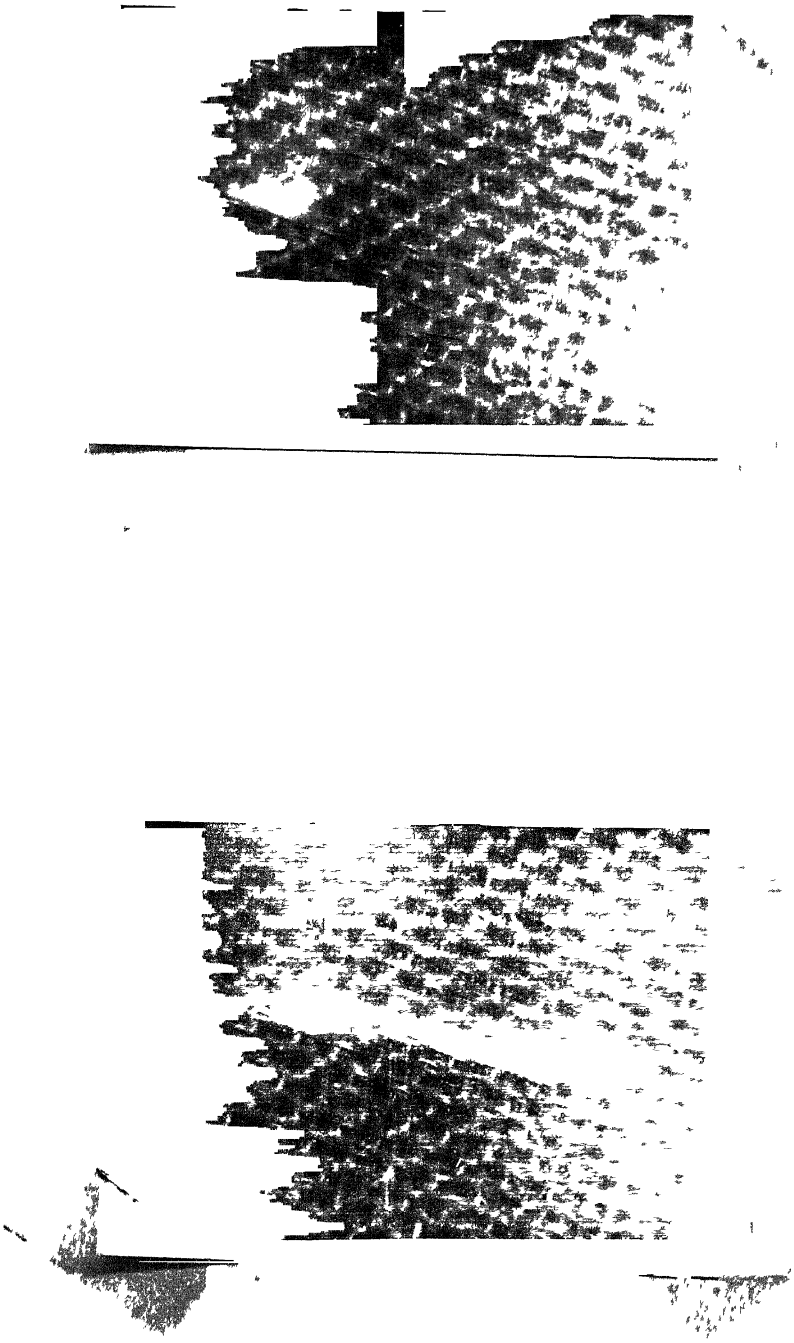


Fig. 24. Photographs showing the stages of crack propagation in the 15° laminate.

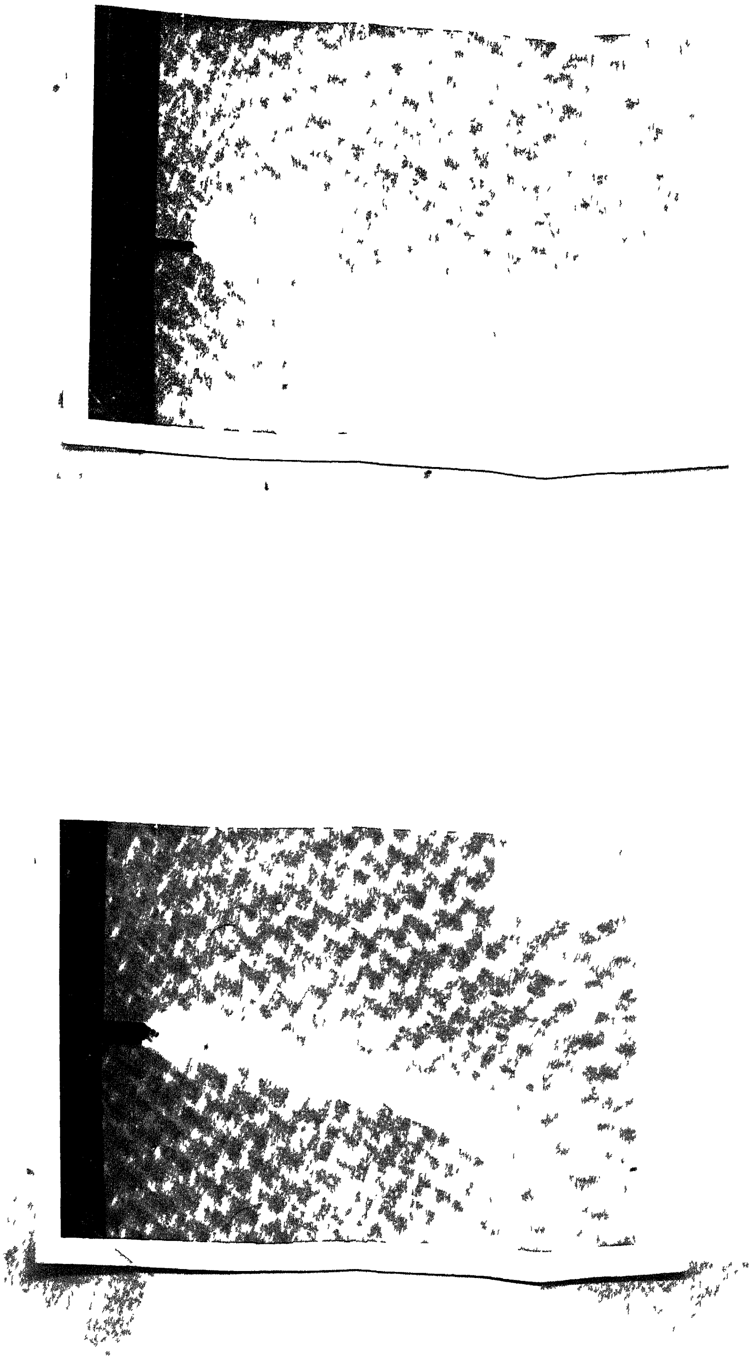


Fig. 25. Photographs showing the stages of crack propagation in the 30° laminate.

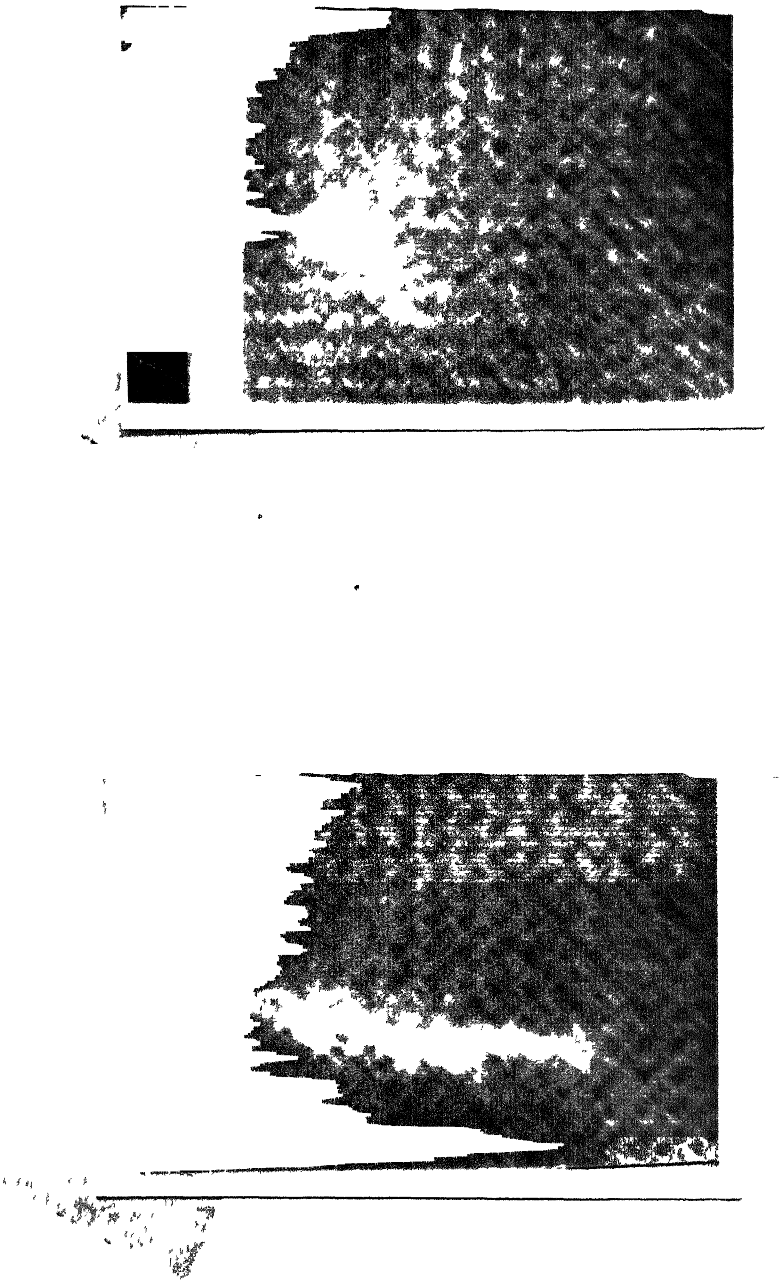


Fig. 26. Photographs showing the stages of crack propagation in the 45° laminate.

fibres near the trailing end of the crack due to high stress concentration,as the fibres there now free of the matrix cannot act in unison. The fracture dominated by tensile failure of the severely stressed interface is further accentuated by the fact that when $\theta = 45^{o}$ and hence when tensile stresses on both the interfaces are equal and they become equally vulnerable, the crack propogates perpendicular to the load axis in a zig-zag manner alternating between the two interface directions in small steps till the end of the sample. Since high amount of whitening is observed in the case of the 45^{o} laminate, not necessarily parallel to the fibres, it is possible that they fail by the shear stress acting upon them as fibres are expected to be quite weak in shear,as the magnitude of the tensile stress on them is least.

Therefore it can be said that whilst the fracture of the 0^{o} laminate is fibre controlled, that of the off-axis ones are dominated by a process of interfacial tensile failure, followed by fibre fracture either by tension or shear depending on the magnitude of such stresses acting upon them. This unique off-axis failure mode of these composites is due to the woven type of reinforcement where the fibres in both the directions act in concert. However the picture would be more clear by supplementary evidence from the nature of the exact stress distribution ahead of the notch by photoelastic techniques and the mode of failure undergone by examination of the fractured surfaces under a scanning electron microscope.

CHAPTER FOUR

CONCLUSIONS AND SCOPE OF FURTHER WORK

4.1 Conclusions:

On the basis of the results reported, the following conclusions can be made:

1) The failure modes of the composite material system described here at fibre orientation parallel to the loading axis is distinctly different from their off-axis failure modes.

2) The crack growth resistance curves of the laminate with $\theta = 0^\circ$ has negative curvatures while those with $\theta = 15^\circ$ 30° and 45° have positive curvatures.

3) The R-curve approach can be expected to yield satisfactory results for composites which are linearly elastic, for which it is possible to determine the parameter $(K_R)_{instability}$ through a compliance matching technique modelling the damage zone propagation as a self similar crack growth. For nonlinear composites this method fails to yield any real solution.

4) The notch sensitivity of the laminate with $\theta = 0^\circ$ is more pronounced than it is for off-axis laminates. The strength reductions due to presence of a notch of the off-axis specimens are almost same within reasonable limits of statistical variations.

5) Both the point and average stress failure criteria for straight cracks proposed by Whitney and Nuismer are valid for the 0^{o} laminate and the off-axis ones in principle. However the values of a_o or d_o proposed by them do not appear to be sacrosanct for all composite material systems, as the values obtained here, for the systems considered by computer best fit are substantially lower than those proposed by them.

6) The stress failure criteria of Whitney and Nuismer for holes do not agree with the results presented in this thesis.

7) The predominant failure mode of the 0^{o} laminate is by fibre fracture perpendicular to the loading axis.

8) The failure mode of the laminates with off-axis orientations are predominantly by interface tensile failure along a susceptible principal fibre direction, the fibres in the other direction rupturing by tensile or shear stress depending on their magnitude which changes with the orientation.

4.2 Suggestions for Further Work

In order to provide the structural designer with simple analytical tools for design with composite materials and the materials designer a better insight so as to be able to develop stronger and tougher composites the following suggestions for further work are proposed.

1) Since the techniques of LEFM developed for metals can only be applied to a limited number of composite material systems, efforts should be directed to predict catastropic fracture in composites of a general nature without resorting to LEFM, but by developing parametric strength models as proposed by Whitney -Nuismer and Pipes et.al.

2) Efforts should be directed to physically interpret the parameters coming into the strength models for composites such as in those above from supplementary evidence by photoelasticity, acoustic emission and quantitative scanning electron fractography etc.

3) Brittle fibre-brittle matrix composite systems by virtue of their energy absorbtion mechanisms during crack growth are expected to outweigh the efficacy of metals at cryogenic temperatures and hence efforts should be directed to study their fracture mechanics under such conditions.

4) Susceptibility of composite materials to stress corrosion cracking should be investigated.

5) Since the effect of shear stress on composite laminates are quite critical, the fracture mechanics in Mode II and III should be thoroughly investigated.

6) To develop better composites, the microfracture processes undergone by different fibre matrix systems should be examined by scanning electron microscope and the results fed back to modify the matrix and the interface so as to enhance their properties.

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